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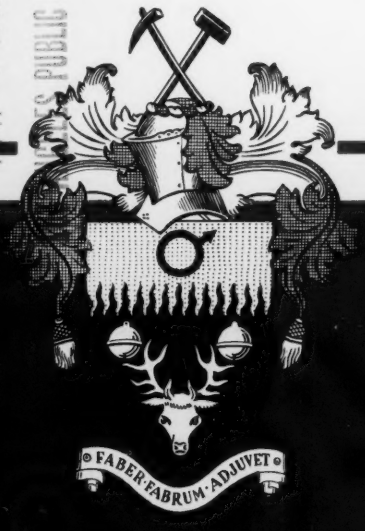

Established 1869 · Incorporated by Royal Charter 1899

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JOURNAL

of

THE IRON AND STEEL INSTITUTE

Volume 158

1948

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PUBLISHED AT THE OFFICES OF THE IRON AND STEEL INSTITUTE
4, GROSVENOR GARDENS, LONDON, S.W.1
1948

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ABBREVIATIONS AND SYMBOLS

Å	Ångstrom unit(s) = 1×10^{-10} m
A.C.	air-cooled ; alternating current
A.H.	air-hardened
amp	ampère(s)
amp.-hr.	ampère-hour(s)
approx.	approximately
at.-%	atomic per cent
at.wt.	atomic weight
atm	atmosphere(s) (pressure)
A.W.G.	American wire-gauge
° Bé	Baumé (degree)
b.h.p.	brake horse-power
B. & S.	Brown and Sharpe (gauge)
B.o.T.	Board of Trade
b.p.	boiling point
B.T.U.	Board of Trade unit(s)
B.Th.U.	British thermal unit(s)
B.W.G.	Birmingham wire-gauge
° C	centigrade (degree)
cal	calorie (-ies)
cc	cubic centimetre(s)
c.d.	current density
c.g.s.	centimetre-gramme-second unit(s)
cm	centimetre(s)
coeff	coefficient(s)
conc	concentrated
const	constant(s)
cu	cubic
cwt	hundredweight(s)
D.C.	direct current
dia	diameter
dil	dilute
dm	decimetre(s)
e.m.f.	electromotive force
e.v.	electron volt(s)
° F	Fahrenheit (degree)
ft	foot, feet
ft.lb.	foot-pound(s)
g	gramme(s)
gal	gallon(s)
H.F.	high-frequency
h.p.	horse-power
h.p.hr.	horse-power-hour(s)
hr	hour(s)
in	inch(es)
in.lb.	inch-pound(s)
I.S.W.G.	Imperial standard wire-gauge
° K	absolute temperature (Kelvin scale)
kg	kilogramme(s)
kg.cal.	kilogramme-calorie(-ies)
kg.m.	kilogramme-metre(s)
km	kilometre(s)
kV	kilovolt(s)
kVA	kilovolt-ampère(s)
kW	kilowatt(s)
kWh	kilowatt-hour(s)
kX	crystal Ångstrom(s) = 1000 Siegbahn X-units
lb	pound(s)
L.F.	low-frequency
M	molar (solution)
m	metre(s)
m.amp.	milliampère(s)
max	maximum
mg	milligramme(s)
min	minimum ; minute(s)
ml	millilitre(s)
mm	millimetre(s)
m.m.f.	magnetomotive force
m.p.	melting point
mV	millivolt(s)
mμ	millimicron = 1×10^{-9} m = 10 Å
N	normal (solution)
N.T.P.	normal temperature and pressure
O.H.	open-hearth ; oil-hardened
O.Q.	oil-quenched
oz	ounce(s)
p.d.	potential difference
*p*H	hydrogen-ion concentration
p.p.m.	parts per million
r.p.m.	revolutions per minute
sec	second(s)
sp.gr.	specific gravity
sq	square
S.W.G.	standard wire-gauge
T	tempered
temp	temperature
V	volt(s)
VA	volt-ampère(s)
W	watt(s)
Wh	watt-hour(s)
W.G.	water-gauge
W.Q.	water-quenched
wt	weight
wt.-%	weight per cent
yd	yard(s)
μ	micron(s) = 1×10^{-6} m
μg	microgramme(s) = 1×10^{-6} g
μμ	1 millionth micron = 1×10^{-12} m = 0·01 Å
Ω	ohm(s)
°	degree (arc or temperature)
′	minute of arc ; foot (feet)
″	second of arc ; inch(es)
$<$	less than
$>$	greater than
$\not<$	not less than
$\not>$	not greater than
$\leqslant$	equal to or less than
$\geqslant$	equal to or greater than
$\neq$	not equal to
$\equiv$	identically equal to
$\approx$	approximately equal to
$\propto$	proportional to

HERBERT HENRY BURTON, F.I.M.

HERBERT HENRY BURTON, who is Chief Metallurgist and Director of Research at the English Steel Corporation, Limited, Sheffield, was educated at Sheffield Royal Grammar School and King Edward VII School and at Sheffield University.

He first joined the staff of the Research Department of Messrs. Cammell, Laird and Company, where he was responsible for a number of investigations, both in the laboratory and in the works. He later became Assistant Superintendent of the Department. After the formation of the English Steel Corporation, Limited, in 1929, Mr. Burton was appointed Senior Assistant to the late Mr. J. H. S. Dickenson, and succeeded him as Chief Metallurgist upon his death in 1934. He was appointed a Special Director of the Company in 1938, and a Director in 1943. Mr. Burton is also a director of Firth-Vickers, Limited, of Darlington Forge, Limited, and of Industrial Steels, Limited.

Mr. Burton was for many years a member of several Joint Committees of The Iron and Steel Institute and the Iron and Steel Industrial Research Council, becoming Chairman of the Ingots Committee and of the Hair-line Crack Sub-Committee on the death of Dr. Swinden. Those Committees now form part of the British Iron and Steel Research Association organization. During the war he was appointed Chairman of the Technical Sub-Committee dealing with gun forgings, and was also a member of the Gun Design Committee, the Metallurgy Committee of the Advisory Council on Scientific Research and Technical Development, the Technical Advisory Committee of the Special and Alloy Steel Committee, and many others. He carried out extensive investigations in connection with steels, for guns, tanks and aircraft armour, and steels for aero-engines. He was a member of the Delegation to the U.S.A. in 1943 in connection with the conservation of alloys in steel manufacture.

For a number of years Mr. Burton has been actively engaged in the investigation of the causes and prevention of hair-line cracks, and the part played by hydrogen in their formation.

Elected a Member of The Iron and Steel Institute in 1935, Mr. Burton became a Member of Council in 1944. In 1947 he was elected a Vice-President.

H. H. Burton, F.I.M.
Vice-President

IRON AND STEEL
INSTITUTE PAPERS

PLANT DESIGN · MANUFACTURE · RESEARCH · DEVELOPMENT

SYMPOSIUM on the PEELING OF WHITEHEART CAST IRON

The occurrence of the defect known as " peel " in whiteheart malleable castings reached serious proportions during the last war, through the failure of track links in military vehicles. The following four papers, introduced by Mr. J. C. W. Humfrey, record the incidence of this defect and show how its causes and the means of elimination were discovered through researches carried out during the war.

Since the findings of the various investigators show some difference of opinion, The Iron and Steel Institute has arranged for the subject to be discussed at a forthcoming special meeting.

INTRODUCTION

By J. C. W. Humfrey, O.B.E., M.Sc., M.Eng., B.A.

THE defect known as " peel " in whiteheart malleable castings occurred spasmodically in pre-war (1939–45 war) days, but its presence could usually be detected in subsequent manufacture and its prevalence never reached serious proportions.

Early in the war extensive use was made of whiteheart malleable castings for various military purposes, one of the main requirements being for the track links of " carrier " vehicles. Peel was not easy to detect by the usual works or official inspection tests, and if peeled links got into service they were a source of considerable danger, their failure in some cases leading to fatal accidents.

The extensive use of whiteheart malleable-iron track links began in 1940, many grey-iron foundries being called in to increase production of castings, and kilns and ovens designed for brick and tile manufacture being adapted for annealing them.

Many failures of the links occurred in service, and my own department of the Ministry of Supply was called in in 1942 to assist the production and inspection departments in overcoming their difficulties. The work was carried out by Dr. Norbury and Mr. Bowden, who were acting as regional officers in the Midlands, where practically all the castings were produced. Examination of a number of failed links returned from service showed that failures were, in general, due to (*a*) inadequate annealing, (*b*) peel, or (*c*) the development of a brittle, coarsely crystalline, structure due to annealing at too high a temperature.

The defects apart from peel could be detected by a non-destructive drop test which was finally applied to 100% of the links, but peel could be detected only by destructive testing, which could be applied to only a relatively small percentage. It therefore became very important to find the true causes of peel and what precautions could be taken entirely to eliminate it. For this purpose both the Cast Iron Control and the officers already mentioned initiated and carried out various experiments in manufacturers' plants, and an extra-mural research contract was placed with the British Cast Iron Research Association. Subsequently a further extra-mural research contract was placed with Mr. A. Preece, who was well known for his work on the scaling of iron and steel. The present symposium collects the results of

Manuscript received 18th September, 1947.

Mr. Humfrey is a Consulting Metallurgist with Messrs. Riley, Harbord and Law, London.

all these researches and shows how the true cause of peel was established and the way pointed for its elimination.

Reports of investigations on peel before the war were not very numerous or illuminating. A brief summary of the chief publications is as follows :

(1) Poole[1] showed that a sulphide concentration was present at the junction of the peel layer with the casting.

(2) Ingall and Field[2] carried out a long investigation. Specimens of white iron of various compositions were annealed together in a laboratory furnace in ore taken from that in use in the works annealing ovens. Various annealing times, temperatures, rates of heating, etc., were used, and the specimens were subsequently tested by examining a bend fracture for separation of a peeled layer. Microscopical examination of peeled samples showed three zones, *viz.* :

(*a*) An outer layer of highly oxidized metal containing free globules of oxide.

(*b*) A zone of practically pure ferrite, becoming more oxidized towards zone (*a*).

(*c*) An inner core of ferrite and pearlite.

Separation of the peel occurred between zones (*a*) and (*b*), and sulphur prints showed that while zone (*a*) was practically free from sulphur, a marked sulphur segregate occurred between zones (*a*) and (*b*). The authors concluded that :

(i) The composition of the annealing ore exercised only a minor influence.

(ii) Increased time of annealing and rate of heating increased the peel effect, which was a maximum at 950° C. and diminished at temperatures above and below.

(iii) Metal containing between 0·20% and 0·40% of sulphur is liable to peeling, and the higher the silicon the greater is the liability to peel.

(3) Oberhoffer[3] differentiated between peel and the scale produced by burning and oxidation. He gives as factors influencing the formation of peel (*a*) the FeO/Fe_2O_3 ratio in the ore, (*b*) the time and temperature of annealing, and (*c*) the size of the ore particles. As regards the chemical composition of the metal, he mentions high sulphur and high silicon content as causing peel, and manganese content as having no influence.

(4) Schuz and Stotz[4] confirmed Ingall and Field's observations on the microstructures found on peeled samples, but also drew attention to the presence of pearlite as well as sulphide at the junction of peel and core.

(5) Taylor,[5] in an investigation into the effect of manganese on the properties of whiteheart malleable castings, showed that sulphur could migrate from the casting to the annealing ore, or *vice versa*. The former happened when the manganese content of the casting was insufficient to combine with all the sulphur as manganese sulphide, and the latter when the manganese was in excess.

(6) Roll[6] confirmed the concentration of sulphur between peel and core. He suggested that peel was caused by the counter-diffusion of carbon outwards from core to surface, and of sulphur and decarburizing gas inwards from the ore.

To summarize these various investigations, the knowledge of peel formation previous to the work described in the present symposium can be said to have consisted of :

(*a*) A fairly clear picture of the microstructure found in a peeled casting, and the definite fact of a sulphide concentration at the peel/core junction.

(*b*) Realization of the possibility of migration of sulphur from ore to casting, or *vice versa*.

(*c*) A theory that high silicon in the white iron was conducive to peel formation.

In pre-war days Cumberland hematite ore (sulphur 0·01%) was used as the decarburizing agent, but supplies of this soon became unavailable and a Riff ore (average sulphur 0·14%) was used as a substitute. It was during this period that the formation of peel became such a serious matter, although even with the use of the substitute ore its occurrence was only serious with certain manufacturers. Since, however, it was evident that high-sulphur ore was one of the main causes of peel formation, the Iron and Steel Control decided to stop supplies of Riff ore and substituted Thabazimbi ore (sulphur 0·01%). This considerably reduced, but did not entirely eliminate, the occurrence of peel.

As will be seen from the following papers, some differences of opinion existed on the influence of ore strength on peel formation. While on the one hand experiments showed that the greatest depth of the peeled layer occurred with the use of strong ores, yet trouble due to peel was considerably less in works where a bright oxidizing atmosphere was maintained in the ovens. The problem appears to be very definitely settled by the work of Preece and Irvine, which shows that while sulphur penetration occurs only in atmospheres high in carbon monoxide, yet its segregation to form a subcutaneous layer is dependent on a corresponding depth of decarburization which must have originally been formed in an oxidizing atmosphere.

The malleablizing process depends on oxidizing conditions in the annealing pots for the major part of the cycle, but gradual weakening of the ore may result in a change to reducing conditions before the cycle is complete, especially if the furnace atmosphere is a reducing one. Given appreciable sulphur content in the ore, such a cycle would appear to be the cause of peel formation. Other factors, such as high sulphur and silicon and low manganese in the white iron, all tend to increase the liability to peel formation.

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THE FORMATION OF PEEL ON WHITEHEART MALLEABLE CAST IRON

By A. Preece, M.Sc., and K. J. Irvine, B.Sc.

SYNOPSIS

The mechanism of peel formation in malleable cast iron has been studied by exposing suitable specimens to mixtures of carbon monoxide, carbon dioxide, and sulphur dioxide. It has been found that the essential conditions for peel formation are the presence of sulphur in the gases and a high carbon monoxide content. The oxide network which is invariably associated with peel formation is caused by preferential oxidation of silicon in the metal, but it is not an essential accompaniment of peel formation, since it is absent when low-silicon materials are treated in a similar manner. In a sufficiently reducing atmosphere absorption and diffusion of sulphur take place, forming a continuous band of iron sulphide beneath the surface of the iron. The depth of this band is directly related to the amount of surface decarburization, and diffusion does not take place until this has proceeded to some extent. The composition of the iron is important only insofar as it governs the rate of decarburization. The results have also shown some interesting features concerned with the diffusion of sulphur in iron.

Introduction

MANUFACTURERS of whiteheart malleable cast iron are troubled by the occasional formation during malleablizing of a film of iron sulphide below the surface of the finished casting. This sulphide film, together with a certain amount of intercrystalline oxidation, with which it is associated, produces a weak surface layer which readily becomes detached from the remainder of the casting on the application of stress. Unfortunately, castings possessing this defect, usually described as "peel," have a perfectly normal appearance, and great care has to be taken to prevent such faulty castings passing into service. The factors which govern peel formation have not been determined, and the present investigation was undertaken to examine the conditions which favour the development of these weak surface structures.

From works observations there is general agreement that the presence of sulphur in the annealing ore plays an important part in peel formation, but manufacturers differ concerning the particular atmosphere in contact with the casting which is most harmful in this respect. Some contend that the more oxidizing the atmosphere the greater the tendency for peel formation, while others hold the opposite view.

It is clear that decarburization of the iron in the malleablizing operation occurs by interaction between the metal and the annealing atmosphere, which consists essentially of carbon monoxide, carbon dioxide, and nitrogen, and usually some gaseous sulphur compound. The balance between the two carbon gases will be controlled by the oxygen pressure of the ore mixture on the one hand and the carbon concentration in the cast iron on the other, and consequently the ratio of carbon monoxide to carbon dioxide will change during the operation.

It was thought, therefore, that a clearer insight into the mechanism might be obtained by malleablizing specimens of white cast iron in mixtures of carbon monoxide, carbon dioxide, and sulphur dioxide, under experimental conditions in which the gas composition was accurately known throughout the process. This method of examination has proved informative and, in addition to elucidating the main problem, has indicated some interesting features concerned with the diffusion of sulphur in iron.

In recent papers by Robiette[1] and Jenkins and Williams,[2] gaseous processes for the production of whiteheart malleable cast iron are described in which partially burnt town's gas is passed over the metal. The advantages claimed are a shorter heating cycle, greater thermal efficiency, and more rigid control. The present authors are in complete agreement with the claims put forward in these papers for the new process, and would add the further advantage that malleablizing in a sulphur-free atmosphere could not give rise to peel formation.

EXPERIMENTAL METHOD

The experimental apparatus is diagrammatically shown in Fig. 1. The two gases, carbon monoxide and carbon dioxide, obtained from cylinders and suitably purified, were passed through calibrated flow gauges, after which they entered a common tube leading to the furnace. Additions of sulphur dioxide could be made to the gas stream immediately before it entered the furnace. In each experiment the specimens were placed in a silica tube, which was then closed and air expelled by allowing the gas mixture to flow through for some minutes before heating in a thermostatically controlled electric furnace.

After treatment each specimen was mounted in Bakelite for surface protection and cut through the centre for microscopic examination. It was thus

Manuscript received 31st May, 1946.

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possible to examine the structure from surface to centre. It was arranged that the periphery of the polished surface of each specimen contained a machined surface, as well as the original cast surface.

When using atmospheres which did not include carbon dioxide the apparatus was modified so that the exit gases could be recirculated, thereby conserving the carbon monoxide. The arrangement is shown in Fig. 1. The exit gases leaving the furnace were passed through a large bottle containing soda asbestos to remove sulphur dioxide and any small quantity of carbon dioxide present. A manometer on the other side of this absorption bottle indicated the pressure inside the apparatus. The gas then passed through a circulating pump provided with a by-pass to control the amount circulated, and finally returned to the carbon monoxide train in front of the constant-pressure mercury seal. A small amount of carbon monoxide was continuously admitted to the circulating system to maintain a pressure slightly above atmospheric, excess gas escaping through the mercury trap at the outlet.

It was found impracticable to use this method of recirculation when using atmospheres containing carbon dioxide. When the exit gases were passed through wash bottles containing weak caustic soda to absorb the carbon dioxide, the solutions had to be so weak to prevent sodium bicarbonate precipitation that the quantities involved became very large, introducing back pressures too great for the circulating pump.

Table I
COMPOSITION OF IRONS USED

Iron	C, %	Si, %	Mn, %	P, %	S, %
High-sulphur white iron (as cast) ...	3·23	0·73	0·22	0·054	0·374
Low-sulphur white iron (as cast) ...	3·25	0·73	0·22	0·054	0·164
Previously malleablized iron ...	...	0·46	0·15	0·061	0·247

EXPERIMENTAL RESULTS

In the first series of experiments, specimens of cast iron were exposed to mixtures of carbon monoxide and carbon dioxide for 100 hr. at 1000° C. The effect of an addition of sulphur to these gas mixtures was also examined in a duplicate set of experiments, in which 0·3% of sulphur dioxide was added to each gas mixture.

Three cast irons were used in each experiment; two were in the cast condition and the third had been malleablized. The analyses of the irons are given in Table I. The first and second were identical except

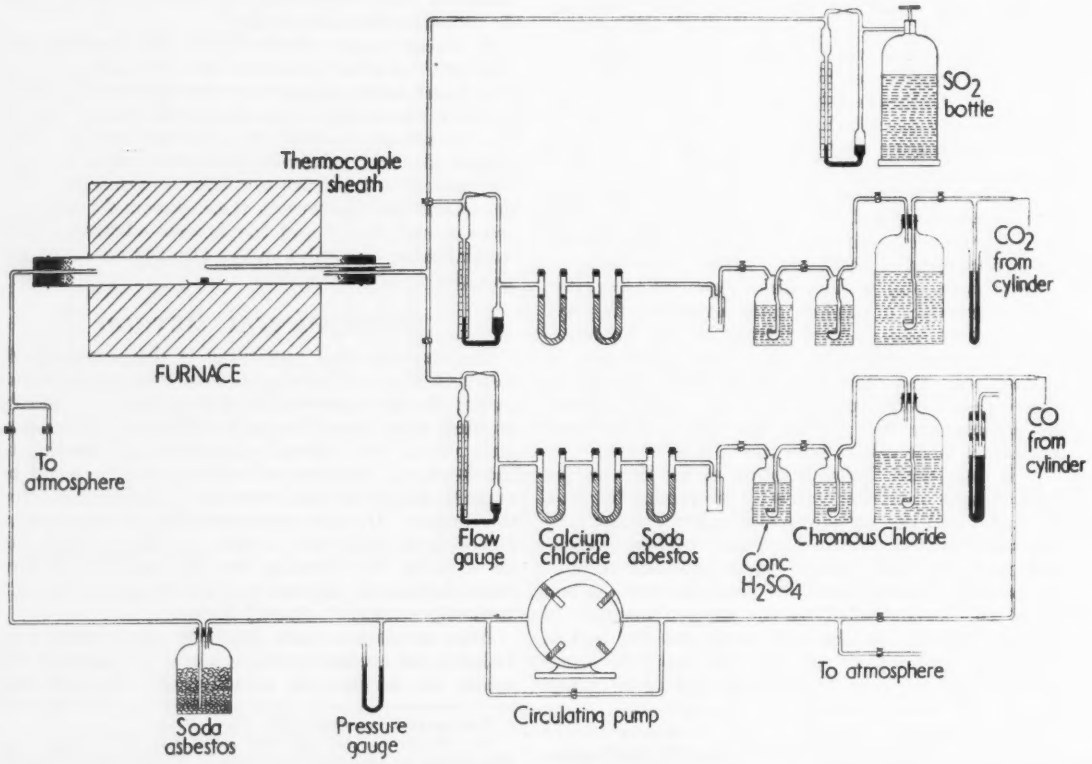

Fig. 1—Diagram of apparatus

for the sulphur content, which is high in one and low in the other. The third iron had been previously malleablized in an industrial furnace and served to indicate the effect of a previous decarburization upon peel formation. The structure of the two white irons was normal, consisting of pearlite and cementite; the malleablized iron consisted of graphite nodules in a matrix of pearlite and ferrite, as shown in Fig. 4.

In examining the structural changes brought about by the treatment described it was found that the original cast surface behaved in a similar manner to the machined surface, except that the former suffered from a varying degree of porosity which resulted in unevenness of attack. Where the surface was sound, however, the depth of the reaction zone was the same as at the machined surface.

It will be convenient to consider the results of each set of experiments separately.

(1) ATMOSPHERES CONSISTING ONLY OF CO + CO_2

Exposure of the cast irons to mixtures of CO + CO_2 at 1000° C. resulted in a slight degree of superficial intercrystalline oxidation and decarburization. It should be stressed that this oxidized layer is very superficial, being only 0·15 mm. deep, and is not important from a practical point of view. The intercrystalline oxidation increased in depth with the carbon dioxide content up to approximately 20% of carbon dioxide, after which there was little further increase; this is shown in Fig. 2. The depth of intercrystalline oxidation was the same for each iron, and the single curve may be taken to represent the behaviour of all three irons.

There was some evidence of decarburization of the irons when heated in carbon monoxide alone, but the carbon content did not fall below the eutectoid composition. On the addition of carbon dioxide, however, areas of ferrite began to appear in the surface layer of metal and increased in extent with the carbon dioxide, as did the intercrystalline oxidation. In atmospheres containing more than 5% of carbon dioxide, decarburization extended beyond the zone of intercrystalline oxidation. Of the two white irons, the one containing the higher sulphur content showed the greater depth of decarburization.

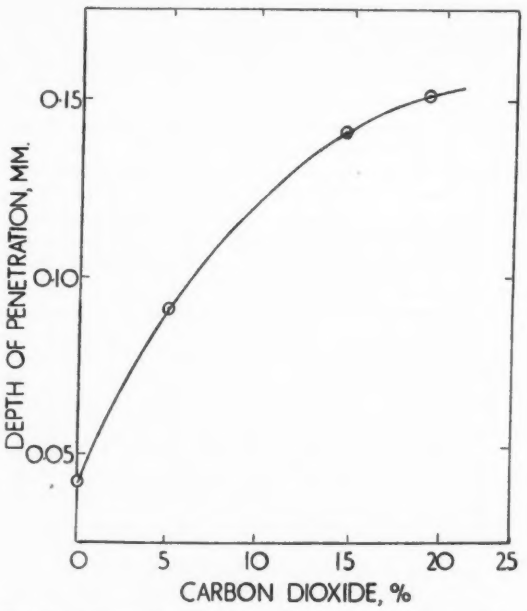

Fig. 2—Effect of increasing carbon dioxide content in the sulphur-free atmosphere upon the depth of intercrystalline oxidation

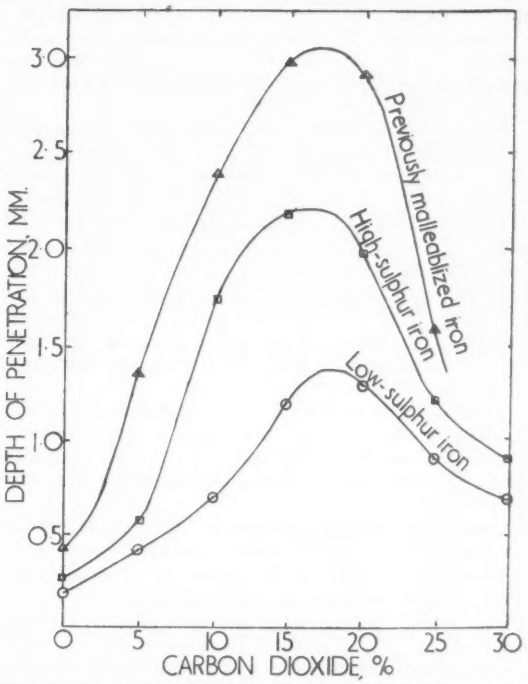

Fig. 3—Effect of increasing carbon dioxide content in the atmosphere containing 0·3% of SO_2 on the depth of intercrystalline oxidation

In atmospheres where appreciable decarburization had occurred there was a continuous band of pearlite at the inner boundary of the zone of intercrystalline oxidation; areas of pearlite were also noted at places inside the oxide network. Between the oxide network and the interior was an almost carbon-free zone, as shown in Fig. 5. This is a good example of the peculiar carbon segregation at the inner limit of the zone of intercrystalline oxidation, and was produced by heating a specimen of white cast iron in an 85%CO/15%CO_2 atmosphere for 100 hr. at 1000° C.

The banded structure illustrated in Fig. 5 results from the sudden fall in silicon content at the inner boundary of the zone of intercrystalline oxidation in which the silicon has been removed by preferential oxidation.

(2) (CO + CO_2) ATMOSPHERES CONTAINING SULPHUR

Exposure of the cast irons to the gas mixtures containing sulphur (0·3% of sulphur dioxide added) resulted in a considerable increase in the depth of

intercrystalline oxidation, which reached a maximum between 15 and 20% of carbon dioxide as can be seen in Fig. 3. Another important feature was the penetration of sulphide into the metal when the carbon dioxide content was below 15%.

Sulphide penetration decreased with increasing carbon dioxide content. It was appreciable up to 10% of carbon dioxide, but with 15% of carbon dioxide hardly any trace of sulphide could be found in the oxide network. With higher concentrations of carbon dioxide, the oxide particles became coarser and more globular, acquiring at the same time the appearance of silicates. As with the sulphur-free gas mixtures, there was a concentration of carbon along the inner boundary of the oxide network, in most cases as a continuous band of pearlite, as illustrated in Fig. 6.

It is clear from Fig. 3 that the original sulphur content of the iron had a marked influence on the depth of decarburization and intercrystalline oxidation. In every case the high-sulphur iron showed the greater attack.

Another important consequence of the presence of sulphur in these gas mixtures was the formation of a thin layer of scale on the surface of the specimens. Only in the absence of carbon dioxide, *i.e.*, with CO + 0·3% of SO_2, were the specimens free from scale. With 5% of carbon dioxide present the scale consisted of iron sulphide, and as the carbon dioxide was further increased the scale became a mixture of iron sulphide and iron oxide. The proportion of iron oxide in the scale increased with the carbon dioxide content until with 30% of carbon dioxide the scale was entirely iron oxide.

The results obtained in these preliminary experiments were interesting in that they indicated the conditions of atmosphere conducive to oxide and sulphide penetration. It also seemed that the structures produced represented some intermediate stage in the production of normal peel. Sulphide penetration had occurred, but as a dispersion of particles throughout the intercrystalline network of oxide rather than as a band, as found in industrial practice. Another important difference was that these specimens were scaled, whereas the surface of commercially malleablized cast iron is free from a scale, even in conditions which promote peel formation.

In each of the foregoing experiments the composition of the atmosphere in contact with the specimen was kept constant during the period of treatment at 1000° C. In the normal malleablizing process, however, there is a continual change in composition, the CO/CO_2 ratio increasing as the annealing ore is progressively reduced by carbon from the cast iron. It seemed likely that this progressive change in composition was responsible for a further reaction in which the structures obtained above are modified to produce the sulphide band found in normal peel.

A further specimen was accordingly heated in a gas stream containing 90% of CO + 10% of CO_2 + 0·3% of SO_2 for 100 hr. at 1000° C., so as to produce a decarburized edge with random distribution of sulphide particles beneath the surface. The same specimen was then further heated for 50 hr. in carbon monoxide at 1000° C. The structure produced after this second treatment is shown in Fig. 7. The continuous band of sulphide is clearly shown and also the clean zone between the band and the surface of the iron. It was also noted that the scale produced in the first treatment had been removed in the second. The sulphide band was somewhat thicker than is normally found in industrial peel, and in the following experiments the sulphur dioxide addition to the gas mixture was reduced to 0·1% in the hope of obtaining structures more closely resembling those found in practice.

The last experiment showed that a change in composition of the atmosphere has a pronounced effect on the final structure. It was therefore decided to try to reproduce the actual atmospheric conditions of the annealing box by exposing specimens of cast iron to a gas mixture in which the carbon dioxide content was progressively varied from 25% to zero as follows:

50 hr. in 75% CO, 25% CO_2 + 0·1% SO_2, followed by
10 hr. in 80% CO, 20% CO_2 + 0·1% SO_2, followed by
10 hr. in 85% CO, 15% CO_2 + 0·1% SO_2, followed by
10 hr. in 90% CO, 10% CO_2 + 0·1% SO_2, followed by
10 hr. in 95% CO, 5% CO_2 + 0·1% SO_2, followed by
10 hr. in 97½% CO, 2½% CO_2 + 0·1% SO_2, followed by
10 hr. in 100% CO, 0% CO_2 + 0·1% SO_2.

In order to follow the changes taking place, a specimen was removed for microscopic examination at each stage in the process. The structural changes are shown in Figs. 8 to 13.

The times of exposure to each atmosphere were chosen quite arbitrarily, the relatively long initial heating time of 50 hr. being merely to give the sufficiently decarburized condition which seems to be necessary before peel formation can occur. The iron used for this and all subsequent experiments contained 3% of carbon, 0·69% of silicon, 0·39% of manganese, and 0·09% of sulphur.

The 50 hours' exposure with 25% of carbon dioxide gave the usual zone of intercrystalline oxidation, but there was no penetration of sulphide into the metal in the following stages and, apart from a slight increase in the amount of intercrystalline oxidation, there was no change in the structure of the specimens until the carbon dioxide content had been reduced to 15%. After treatment in this atmosphere the specimens were covered with an oxide–sulphide scale and there were also minute particles of oxide and sulphide at isolated regions along the inner limit of the intercrystalline oxidation, as shown in Fig. 10. With still less carbon dioxide the sulphide particles gradually assumed the typical band formation found in peeled castings. The evidence of a sulphide band was slight with 10% of carbon dioxide, but was clearly defined after the 10 hours' treatment with only 5% of carbon dioxide.

A further change was observed when the carbon dioxide content was brought down to 2·5%. Here there was a large increase in the amount of sulphide taken up by the metal, and the final treatment in CO + SO_2 still further consolidated the sulphide band. The position of the sulphide band coincided with the limit of surface decarburization, and the influence of the latter in governing the depth of penetration of the sulphide band was confirmed in the following manner.

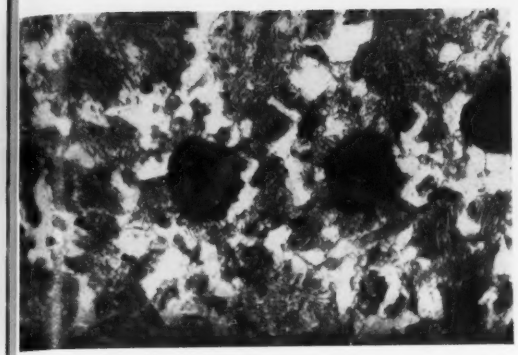

Fig. 4—Initial structure of previously malleabilized iron, showing extent of decarburization × 200

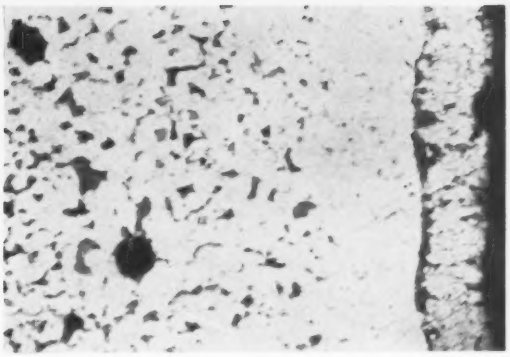

Fig. 5—White cast iron treated in 85% CO + 15% CO_2 for 100 hr. at 1000° C., showing extent of decarburization and concentration of pearlite near surface × 75

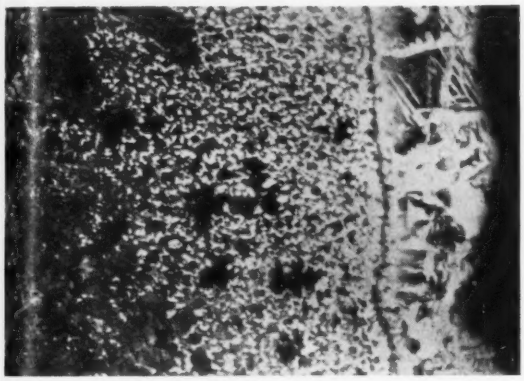

Fig. 6—White cast iron treated in 70% CO + 30% CO_2 for 100 hr. at 1000° C., showing concentration of carbon at inner limit of intercrystalline oxidation × 40

Fig. 7—Sulphide band produced after treating in CO at 1000° C. a specimen previously treated in 90% CO + 10% CO_2 + 0·3% SO_2 × 100

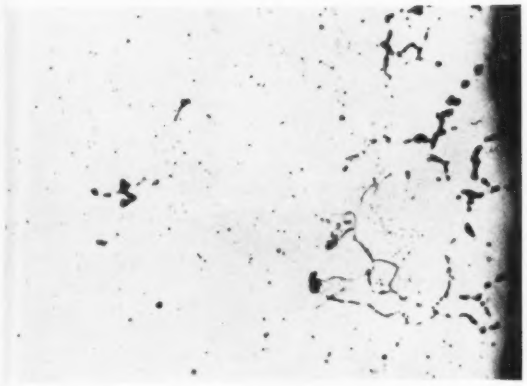

Fig. 8—White cast iron after 50 hours' treatment in 75% CO + 25% CO_2 + 0·3% SO_2. The gradual development of peel is shown in Figs. 9 to 13 × 160

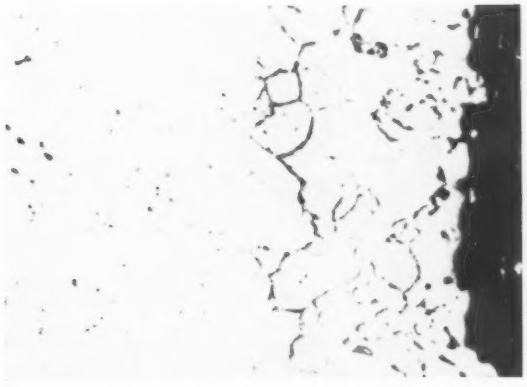

Fig. 9—Same specimen as in Fig. 8 after further treatment for 10 hr. in 80% CO + 20% CO_2 + 0·3% SO_2 × 160

(Figs. 4, 5, and 6 etched with 2½% nital)

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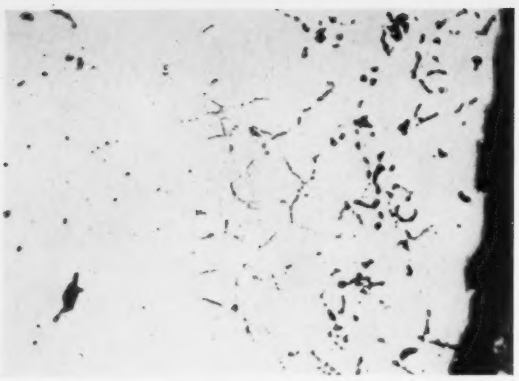

Fig. 10—Same specimen as in Fig. 9 after further treatment for 10 hr. in 85% CO + 15% CO_2 + 0·1% SO_2, showing fine particles of oxide and sulphide × 160

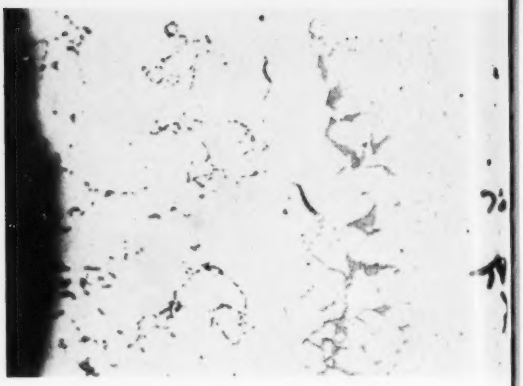

Fig. 11—Same specimen as in Fig. 10 after further treatment for 10 hr. in 90% CO × 10% CO_2 + 0·1% SO_2, showing sulphide band building-up beneath zone of intercrystalline oxidation × 160

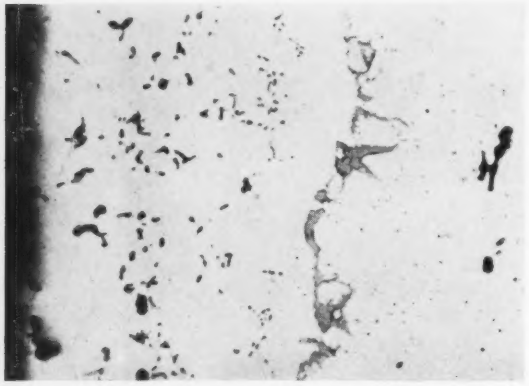

Fig. 12—Same specimen as in Fig. 11, after further treatment for 10 hr. in 95% CO + 5% CO_2 + 0·1% SO_2 × 160

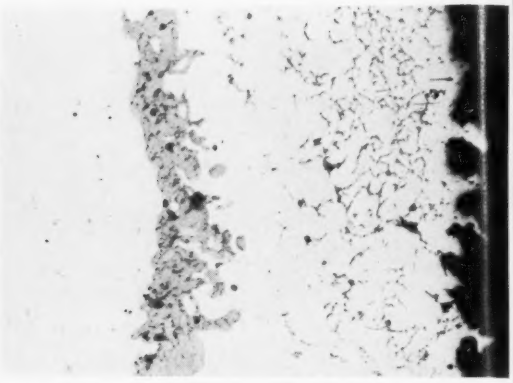

Fig. 13—Same specimen as in Fig. 12, after further treatment for 10 hr. in 97½% CO + 2½% CO_2 + 0·1% SO_2. The longer treatment has considerably consolidated the sulphide band × 160

Fig. 14—A 0·75% carbon steel coated with iron sulphide paste and treated in CO for 20 hr. at 1000° C. × 40

Fig. 15—Sulphur print of peeled track link, showing concentration of sulphide

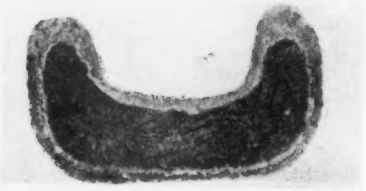

Fig. 16—Sulphur print of the track link shown in Fig. 15, after 50 hr. treatment in 80% CO + 20% CO_2. The sulphide has been replaced by oxide

Several specimens were given an initial treatment in a 90%CO/10%CO_2/0·3%SO_2 atmosphere, which gave a decarburized zone approximately 0·6 mm. deep, together with a slight degree of sulphide penetration and a scale of oxide and sulphide on the surface of the metal. These specimens were then heated at 1000° C. in carbon monoxide for varying lengths of time. After only 2 hr. the sulphide had commenced to move into the metal, and at the end of 5 hr. a definite band of sulphide had formed beneath the surface. After 10 hr. this sulphide band had diffused 0·4 mm. into the iron and there was no trace of sulphide scale on the surface of the specimen. Longer annealing times increased the penetration of the band very slightly, and after 50 hr. its depth inside the metal was 0·5–0·6 mm. Ample evidence was also obtained in the earlier experiments in this work that a certain amount of decarburization was necessary before sulphide penetration became possible.

Since oxidation of carbon and silicon occurred together a concentration gradient of silicon as well as carbon existed in the surface layers of metal. Although the separate effects of carbon and silicon on the diffusion of the sulphide have not been ascertained, the combined effect of these two concentration gradients is to decrease the rate of diffusion of the sulphide as it moves further away from the surface. Thus, although the sulphide exists originally as separate particles, the inward diffusion along the grain boundaries must inevitably lead to the formation of a continuous band.

The importance of a sulphide scale in promoting penetration of sulphide was demonstrated by covering specimens of cast iron with a thin coating of iron sulphide or mixtures of iron oxide and iron sulphide and treating these specimens in CO/CO_2 atmospheres. The sulphides were finely ground and mixed with Bakelite varnish and acetone to a suitable consistency. A thin layer could then be painted on the metal and hardened before placing in the furnace. Artificial scales put on the metal in this way behaved in a similar manner to scales produced by heating the cast iron in the appropriate $CO + CO_2 + SO_2$ mixtures.

Specimens which had been previously malleablized to give differing depths of decarburization (but no sulphide penetration) were painted with iron sulphide before heating in carbon monoxide. Here again it was found that the depth of penetration of the sulphide band was governed by the depth of decarburization. In a specimen with a particularly deep decarburized zone, the diffusion in the region of very low carbon content followed the ordinary diffusion law with regard to time. A similar experiment using a specimen of 0·75% carbon steel instead of the cast iron gave a penetration of sulphide as shown in Fig. 14. There was no evidence of a sulphide band, owing presumably to the absence of a carbon concentration gradient. It was noted that as the sulphide moved into the metal it left a cleaner iron in its wake. It seemed to dissolve and carry with it particles of manganese sulphide and other impurities.

Specimens with an artificial coating of iron sulphide and treated in CO/CO_2 atmospheres behaved in a similar manner to specimens on which a sulphide scale had been produced by treatment in the appropriate atmosphere. For example, with 20% of carbon dioxide present the sulphide coating was removed, leaving a powdery deposit, and sulphide penetration from the coating did not occur until the carbon dioxide content had been brought down to below 15%. With lower concentrations of carbon dioxide, penetration was deeper, and the formation of a sulphide band occurred as previously described.

Further experiments in which specimens were coated with iron-oxide/iron-sulphide mixtures and subsequently heated in carbon monoxide showed that sulphide penetration occurred only when the coating contained more than 50% of iron sulphide. This result is important, since it illustrates the importance of the composition of the scale in determining whether sulphide penetration will occur during the final stages of the malleablizing process. This was clearly demonstrated by the following experiments.

A specimen of white cast iron was heated in an 80%CO/20%CO_2/0·3%SO_2 atmosphere for 100 hr. at 1000° C. to produce a scale consisting mainly of iron oxide with a small proportion of iron sulphide. There was also a little intercrystalline oxidation at the surface. After further heating in carbon monoxide the depth of oxidation had increased slightly, but there was no penetration of sulphide. This result is similar to that obtained with an iron-oxide/iron-sulphide coating containing a preponderance of iron oxide.

This experiment was repeated with the initial annealing 75%CO/15%CO_2/0·3%SO_2 atmosphere. This produced a scale consisting approximately of equal parts of iron oxide and iron sulphide, and there was also some intercrystalline oxidation. After the additional heating in carbon monoxide, a band of sulphide had formed 0·7 mm. below the surface.

A third experiment, in which an initial atmosphere of 90% of CO, 10% of CO_2, and 0·3% of SO_2 was used, gave a scale consisting chiefly of iron sulphide with a smaller quantity of iron oxide. There was also a little penetration of sulphide, which occurred as particles associated with the intercrystalline network. The subsequent treatment in carbon monoxide produced a continuous band of sulphide 0·4 mm. below the surface, and the amount of sulphide was considerably more than in the previous experiment.

The difference in the depth of penetration of the bands in the last two experiments, *i.e.*, 0·7 mm. and 0·4 mm., is explained by reference to previous results, where it was shown that in $CO + CO_2$ atmospheres containing sulphur, decarburization is greater when 15% of CO_2 is present than with only 10% of CO_2, and, as shown previously, the depth of decarburization also controls the depth of sulphide penetration.

It has been indicated earlier that sulphur absorption direct from the annealing at osphere can occur only when the carbon dioxide content is below approximately 15%. This concentration, however, cannot be regarded as the upper limit for sulphide penetration, since an atmosphere containing 15% of carbon dioxide can produce a scale containing iron sulphide which will itself induce sulphide penetration if the atmosphere should become richer in carbon monoxide. Thus the safe working limit would appear to be about 20% of carbon dioxide.

Experiments in which irons of different compositions were treated in the appropriate atmospheres have shown that high silicon encourages peel formation, while manganese has the opposite effect. The prevention of peel formation by the judicious combination of these two elements is, however, inadvisable, since they both have an important influence on graphitization. A more obvious remedy would be the use of sufficient virgin ore in the malleablizing process.

Though it is of no practical significance, it may be pointed out that the sulphide band in peeled castings can be removed by annealing for prolonged periods in an 80%CO/20%CO_2 atmosphere, but this does not produce a sound casting, as the iron sulphide is converted into a band of iron oxide. Sulphur prints of a peeled section before and after this treatment are shown in Figs. 15 and 16. The black line in Fig. 15, indicating the sulphide, appears as a white line in Fig. 16, owing to the replacement of this iron sulphide band by iron oxide. The possibility of building-up an oxide layer well below the surface of the iron by the diffusion of oxygen inwards from the surface would appear to present a new problem in the study of diffusion processes.

CONSIDERATION OF RESULTS

The results obtained in the present investigation have indicated the conditions which give rise to peel formation and the mechanism concerned. The necessary conditions are:

(1) The presence of sulphur in the atmosphere surrounding the cast iron.

(2) A highly reducing atmosphere (a high CO/CO_2 ratio).

(3) Partial decarburization of the white iron.

Absorption of the sulphur by the metal may take place either directly from the atmosphere or from an oxide–sulphide scale formed on the surface of the metal during the malleablizing process. The oxide–sulphide scale can be formed in atmospheres high in carbon dioxide, but it is not until the atmosphere contains less than 15% of carbon dioxide that the sulphur diffuses into the metal. Thus the scale will form during the early stages of the malleablizing process, and later, when the atmosphere becomes highly reducing, the sulphur moves inwards from the surface and forms a band of sulphide at the inner limit of the decarburized zone.

The depth of this band is therefore governed by the extent of surface decarburization, and factors which influence this will in turn govern the thickness of the peel. Thus an original high sulphur or high silicon content in the iron will tend to give a thick peel, by virtue of the fact that they accentuate decarburization.

The preferential oxidation of the silicon to give an intercrystalline network of oxide is not an essential accompaniment to sulphur penetration, as has been indicated by the experiments on plain carbon steel. However, since malleablized cast iron always contains an appreciable silicon content, peeling is invariably associated with the oxide network.

It has been clearly demonstrated that as peel formation requires an extremely high CO/CO_2 ratio in the annealing atmosphere, a simple method of prevention is to be found in the use of ore mixtures containing adequate proportions of ferric oxide.

It follows from this work that the gaseous process for malleablizing will be free from troubles due to peel formation, since the atmosphere in contact with the iron is controlled throughout the operation and will be of constant composition and free from sulphur.

The precise action of carbon monoxide in promoting sulphur diffusion in iron as described in this investigation presents a new and very interesting problem. Much work has already been carried out on this more fundamental question, and the results of this further study have recently been published.[3]*

Acknowledgments

The authors wish to thank the Chief Scientific Officer, Ministry of Supply, for permission to publish this work. They would also like to express thanks to Mr. J. C. W. Humfrey and Dr. A. L. Norbury (Ministry of Supply, Superintendent Technical Application of Materials) for helpful discussion, and to the staff of the British Cast Iron Research Association and Mr. G. R. Webster (Britannia Iron and Steel Works, Ltd.) for supply of materials. Thanks are also due to Professor Townend for the provision of facilities.

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3. A. Preece and K. J. Irvine: *Journal of The Iron and Steel Institute*, 1947, vol. 157, Nov., pp. 336–343.

* The synopsis of the paper "Sulphide Penetration in the Carburization of Steel," by A. Preece and K. J. Irvine, is as follows:

The mechanism of sulphide penetration into steel during carburization has been demonstrated by experiments in which suitably coated specimens of Armco iron have been exposed to carburizing mixtures of carbon monoxide and carbon dioxide. By adjusting the gas mixture in contact with specimens, deep-seated inclusions of iron sulphide have been converted into iron oxide. Iron sulphide inclusions have also been converted into manganese sulphide by prolonged treatment at 1000° C. The results have also shown interrelated effects between carbon, sulphur, and oxygen which may be of significance in the segregation and homogenization of steel castings.

PEEL ON WHITEHEART MALLEABLE CAST IRON

By J. S. Bowden, B.Met., F.I.M.

SYNOPSIS

In the early stages of the recent war peeling troubles were prevalent in whiteheart malleable-iron foundries. The defect was observed in all types of castings and was responsible for a considerable number of failures in service, carrier-track links being particularly affected. Tests were devised to detect peel on track links and on other castings which were to be put into service without machining. Works experiments, instituted to investigate the conditions under which peeling takes place, are described. A theory is advanced that peel formation occurs by preferential oxidation of the silicon in the metal to form fayalite when the CO/CO_2 ratio in the malleablizing atmosphere is abnormally high. Other investigators have shown that peel can be produced by annealing in sulphurous ore, and Preece produced heavy peel by annealing in 100% CO + 0·3% SO_2. It is also demonstrated that peel can be produced on silicon–manganese steel.

DESCRIPTION OF PEEL

PEELING, a defect which has been known to malleable-iron founders and machinists for many years, was prevalent on whiteheart malleable-iron castings during the recent war and was responsible for many failures, particularly of carrier-track links. Peel, which develops during the malleablizing process, is totally different from the more common scale[1] formed when ferrous metals are heated in air at temperatures in excess of 600° C. Pfeil[2] has shown that three distinct layers, each consisting of ferric oxide (Fe_2O_3) and ferrous oxide (FeO) may be distinguished in scale. In the outer layer those oxides are present in the approximate proportions of 21% of FeO to 79% of Fe_2O_3; in the middle layer the proportions are 71% of FeO to 29% of Fe_2O_3, and in the innermost layer they are about 82% of FeO to 18% of Fe_2O_3. The outer and middle layers of scale are dense and break with a crystalline to conchoidal fracture, and the inner layer is weak, porous, and friable. An example of scale on whiteheart malleable cast iron is shown in Fig. 5.

Peel is characterized by the formation of a partially oxidized surface layer of metal of low mechanical properties, in conjunction with a brittle iron sulphide-rich film at the peel/metal interface, as shown in Fig. 6. The outer (peeled) layer, which is largely ferritic with a considerable admixture of non-metallic inclusions such as coalesced oxides, silicates, and sulphides, tends to separate from the casting when lightly stressed. Peel buckles readily in compression, and separation of the layers owing to shock or mechanical deformation leads to an appreciable reduction in strength and effective thickness (*see* Fig. 4).

OCCURRENCE OF PEEL

As the industry was largely concerned with the production of track links, this report deals with investigations carried out upon castings of this type. It was observed that peel was most frequently present upon the horns. pinhole-walls, and other thin sections of the castings, as shown in Fig. 1, and that it tended to be thicker on the inner walls of pinholes than on the outside (*see* Fig. 1 (*c*)) and in the centre of the lugs than at the ends. The occurrence of peel was irregular, but it was encountered more frequently in certain annealing plants than in others. In many cases peeling was extremely localized and varied in degree and extent even on different parts of the same

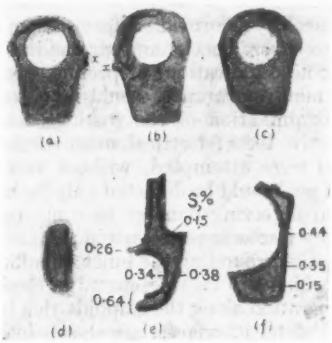

Fig. 1—Sulphur prints of sections of peeled track link. Sulphur pick-up in (*e*) and (*f*)

casting; Fig. 2 shows sulphur prints of different sections of a badly peeled track link which failed through the male lugs in service. In this particular case the defect was confined to one side of the casting, the driving boss and three-hole side (Fig. 2 (*f*)) being completely free from peel.

An appreciable proportion of track-link production was found to be affected by peel in varying degree or to display evidence of incipient peeling. Link failures owing to peeling resulted in broken tracks and were

Manuscript received 10th February, 1947.

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Fig. 2—Sulphur prints of sections cut from a single peeled track link

responsible for a number of accidents ; an example of this type of failure is shown in Fig. 3. Peeling was also encountered in pipe fittings, bracket couplings, rocket grids, and the thin sections of other castings.

TESTS TO DETECT PEEL

Machining operations such as thread-cutting or milling revealed the presence of peel in many types of casting, as the defective layer was exfoliated by the cutting tools, but track links presented a special problem, as the castings—which were seldom machined or mechanically deformed before assembly—were subjected to severe shocks and strains in service.

No distinctive indications of peeling, apart from a dull grey mat appearance, could be observed from a surface examination of the castings, and various non-destructive tests (electrical, magnetic, supersonic, and X-ray) were attempted, without success. The presence of peel could be detected only by mechanical deformation involving damage to some part of the casting or by fracturing the casting ; the well-defined layers of peel surrounding the inner metallic core of a track link which had been fractured in the bend test, and the separation along the sulphide-rich boundaries at the peel/metal interfaces, are shown in Fig. 4.

Tests were devised to detect the presence of peel on carrier-track links whilst restricting the superficial damage to relatively unimportant parts of the casting. Peel was revealed (i) by a sharp blow on the tip of the horn with a 2-lb. hammer or the sharp point of a fettler's pick (Fig. 3, top left), (ii) by vibration set up by the riveting machine during the link-assembly process (Fig. 3, right foreground), (iii) by squeezing the tips of the horns in a vice, and (iv) by milling the ends of the lugs. The last-named operation, which is for the purpose of ensuring interchangeability of links, is not carried out in all works.

Test-pegs $\frac{1}{4}$ in. in dia. and $\frac{1}{2}$ in. long, cast on each link and broken off after annealing, provided a satisfactory means of detecting peel and formed a part of the routine inspection in plants which were experiencing peeling troubles, but this test was far too slow and laborious for general application to large-scale production.

The drop test gave indications of peeling, if present on the central (driving) boss of the link, in the form of fine surface cracks but did not provide a reliable means of detection.

Close works inspection along the above lines did much to eliminate peeled castings from production, and peel-detection charts, circulated to ordnance depots and field units by the Directorate of Mechanical Engineering of the Ministry of Supply (M.E. 7), reduced the number of peeled links in the service. Modification of design (namely, thickening the male-lug walls and eliminating the recess shown in Fig. 3) also reduced the risk of failure, but it was considered

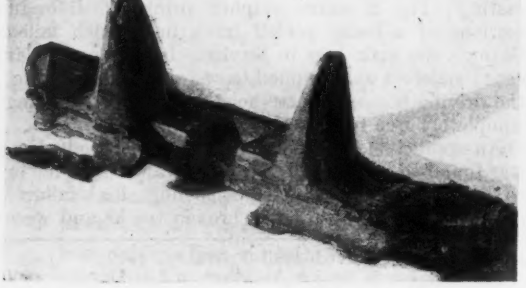

Fig. 3—Peeled track link which failed after 600 miles in service. Peel confirmed after failure by hammer-test on horn

Fig. 4—Fracture of peeled track link

advisable to discover the causes of peeling and methods of eliminating the defect.

INVESTIGATION OF THE CAUSES OF PEELING

(a) Works Experiments

One of the first experiments was to cast track links in widely different compositions of metal (silicon 0·5–1·2%, manganese 0·1–0·4%, sulphur 0·1–0·3%) and to anneal them under ordinary commercial conditions in different ovens. No appreciable peel was produced. Further experiments with links of similar extreme compositions (silicon 0·5–1·5%, manganese 0·1–0·5%, sulphur 0·1–0·3%), which were annealed in ore mixtures of varying strengths at different temperatures for long and short periods in the shot-blasted, tumbled, and as-cast conditions, also failed to produce appreciable peel. Links annealed in ore to which 0·5% of sulphur in the form of ferrous sulphide or pyrites had been added showed evidence of intercrystalline sulphide penetration, but the subcutaneous band of sulphide-rich inclusions characteristic of peel had not been formed.

About this time Mr. Proud, of Messrs. Bagshawe and Co., Ltd., Dunstable, observed that local spots of deep peel were associated with the adherence of mineral particles, which evolved hydrogen sulphide when treated with acid. Experiments were made in which particles of minerals likely to occur in hematite ores were placed in contact with the castings during annealing. No evidence of peel was observed in these experiments, but pyrites and iron sulphide—particularly when in association with calcareous minerals—caused considerable intercrystalline sulphide penetration.

Examination of links submitted for acceptance tests showed that peel tended to be thicker on the inner walls of pinholes than on the outside, as shown in Figs. 7 and 8, and suggested that some compound in the core may have been responsible for peeling. Certain core-binders made with sulphite lye were found to contain up to 0·176% of sulphur. Attempts to produce peel by dusting the oil–sand cores and the mould faces with iron sulphide, pyrites, and sodium sulphite in order to simulate sulphur pick-up during casting were not successful.

Observations in a large number of works showed that peeling was most prevalent in ovens which were operated with a smoky flame and a stagnant atmosphere and that the spent annealing ores from these ovens were very weak, *i.e.*, low in oxygen content and bulk density. It was then realized that all the experiments had been carried out in ovens which were operated with a bright clear flame and a buoyant atmosphere and in which the spent ores were relatively strong, as shown in Table I, and that the CO/CO_2 ratio in the annealing atmosphere might be of significance. Further track links were annealed under the same conditions as before, *i.e.*, in ore mixtures to which 0·5% of sulphur as ferrous sulphur had been added, but pieces of graphite rod were placed inside the pinholes in order to promote localized concentrations of carbon monoxide.

Thick peel (about 0·06 in. deep) was produced on the inside of the hole (Fig. 9) whilst the outer surfaces were comparatively free from the defect (Fig. 10). Similar results were obtained when other track links were annealed with graphite inside the pinholes, in normal annealing-ore mixtures, in commercial furnaces, and in an electric muffle furnace.

Other links were annealed under normal commercial conditions with the lugs of one link nesting into the recesses of a second link so as to protect the gas generated (carbon monoxide) inside the pinhole

Table I

TYPICAL ANALYSES OF ANNEALING ORES

	Oxygen as Fe_2O_3 and FeO*		Fe_2O_3, %	FeO, %	Metallic Fe, %	SiO_2, %	S, %
	lb./cu. ft.	%					
New Ores:							
Cumberland	...	24·5	81·0	1·0	Nil	14·6	0·01
	34·6	26·0	86·6	Nil	Nil	9·3	0·029
Riff	...	26·6	86·3	Nil	Nil	4·6	0·14
	37·1	27·3	89·9	1·29	Nil	4·1	0·09
Thabazimbi	43·1	28·7	95·7	Nil	Nil	2·49	0·01
	...	28·8	95·8	0·29	Nil	3·75	0·01
Spent Ores:							
Free from peeling troubles ...	22·0	18·3	8·65	71·4	6·7	8·7	0·24
	21·7	17·4	14·28	59·5	13·7	8·4	0·23
	20·4	15·1	9·40	56·0	21·2	7·0	0·23
	16·3	13·9	7·2	53·5	23·5	11·0	0·14
	10·8	8·6	15·9	17·5	56·4	7·5	0·39
Peeling prevalent	8·7	7·9	5·8	28·2	56·7	7·0	0·27
	5·8	5·5	1·8	22·5	60·3	10·4	0·66
	4·45	3·8	0·25	17·03	76·6	7·8	0·09
	3·6	3·4	2·75	9·34	83·8	4·7	0·08
	3·5	3·4	1·2	13·6	70·7	9·0	0·41

* Calculated

from easy access to and oxidation by the annealing ore. Peel formed inside the male-lug pinholes but was not detected in other parts of the casting.

Considerable thicknesses of peel (up to 0·08 in.) were produced on castings malleablized in a mixture of spent ore (sulphur 0·2%) and 20% of coke.

It was concluded that since sulphur is always present in the annealing ores (sulphur 0·2% or over) and in the furnace gases, relatively reducing atmospheres inside the annealing pans were an essential factor in the production of peel. These reducing atmospheres could be due to the use of weak ore mixtures or to the maintenance of high CO/CO_2 ratios in the furnace atmosphere.

The results of experiments in which samples of new Thabazimbi hematite ore and a very weak spent ore were heated in small well-decarburized pots are given in Table II and showed that the furnace gases penetrate into the interiors of the annealing pots and play an important part in the malleablizing process.

A series of experiments initiated by Messrs. Bagshawe and Co., Ltd., the Inspectorate of Fighting Vehicles, the Directorate of Tank Design, the Iron and Steel Control, and the Superintendent of the Technical Application of Metals, and carried out by Mr. Proud, of Messrs. Bagshawe and Co., Ltd., and Mr. A. A. Timmins of the Iron and Steel Control, had led to the conclusion that peel formation was associated with the nature of the spent annealing ore. These experiments, which involved an interchange of castings and annealing ore between a number of firms, were conducted before the significance of the oxygen content of the spent ore had been appreciated, and showed that peel was formed in annealing ores which had a high sulphur content.

The Iron and Steel Control, acting upon the evidence of these experiments and the impression, prevalent in the industry, that sulphurous annealing ores were the sole cause of peeling troubles, issued supplies of low-sulphur Thabazimbi ore (sulphur 0·01%) to replace the Riff ores (sulphur 0·14%) which were in use as a war-time substitute for Cumberland hematite (sulphur 0·01%). This change had a beneficial influence, but, in support of the theory that peel formation is primarily due to the use of insufficiently oxidizing atmospheres, it may be argued that Thabazimbi ore is a very dense massive hematite of low porosity, that ore mixtures—which are made up by volume—would contain more than the customary amounts of oxygen (*see* Table I), and that several firms which maintained a strong ore mixture, a buoyant furnace atmosphere, and a strong spent ore experienced little or no trouble from peel whilst using Riff ores.

(b) Works Experience

At an annealing plant in which many of the unsuccessful attempts to produce peel experimentally had been made, trouble was experienced with a pulverized-fuel burner, resulting in the formation of a pile of unburnt coal dust in one of the annealing ovens. The contents of this oven were examined very thoroughly, but peeled castings were found only in those cans which were adjacent to or immediately over the pile of smouldering fuel, that is, where the CO/CO_2 ratio was abnormally high.

Table II

ANALYSES OF ANNEALING ORES HEATED IN SEALED AND LUTED PANS OUT OF CONTACT WITH WHITE IRON

	Annealing Cycles	Fe_2O_3, %	FeO, %	Metallic Fe, %	SiO_2, %	S, %	Oxygen as Fe_2O_3 and FeO,* %
Original Spent Ore:		1·81	22·5	60·3	10·4	0·66	5·5
Heated in oven *A*	1	0·18	21·8	65·4	9·5	0·59	4·85
Heated in oven *B*	1	11·8	54·8	11·8	7·05	0·33	15·60
Heated in oven *C*	1	42·17	47·44	1·12	...	0·48	23·1
Heated in oven *D*	1	27·68	50·66	6·03	8·85	0·50	19·45
" " " "	2	84·7	4·1	0·67	...	0·158	26·3
New Thabazimbi Ore:		95·84	0·29	Nil	3·75	0·01	28·81
Heated in oven *A*†	1	11·8	74·1	6·7	6·5	0·175	19·84
Heated in oven *B*	1	17·9	72·6	17·9	5·24	...	21·37
Heated in oven *C*	1	85·46	8·9	Nil	...	0·08	26·2
Heated in oven *D*	1	63·10	26·9	0·73	6·90	0·15	24·85
" " " "	2	86·6	0·72	0·56	7·80	0·033	26·1

*The data for oxygen as Fe_2O_3 and FeO are calculated figures. †Peeling prevalent in oven *A*

Notes:
Oven *A*: Coal-fired oven with smoky flame and stagnant atmosphere
Oven *B*: Producer-gas-fired oven
Oven *C*: Producer-gas-fired tunnel furnace
Oven *D*: Pulverized-fuel-fired oven with clear flame and buoyant atmosphere

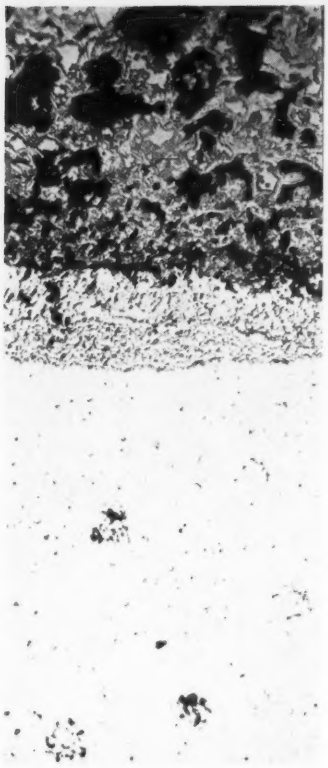

Fig. 5—Scale on whiteheart malleable cast iron (note absence of carbide and sulphur-rich layer at scale/metal interface) × 60

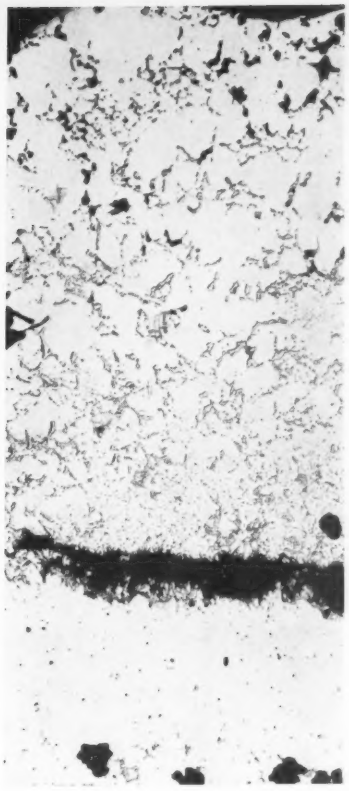

Fig. 6—Peel on whiteheart malleable cast iron track link shown in Fig. 4 × 60

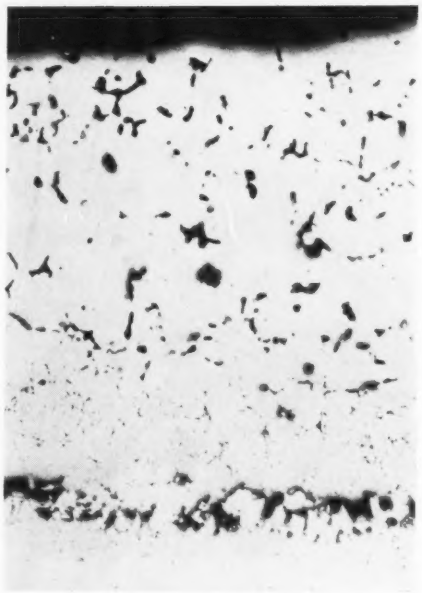

Fig. 7—Peel on inner wall of a track-link pinhole. Unetched × 60

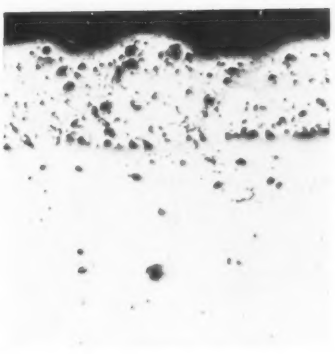

Fig. 8—Peel on outer surface of the pinhole wall shown in Fig. 7. Unetched × 60

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[*To face p.* 12

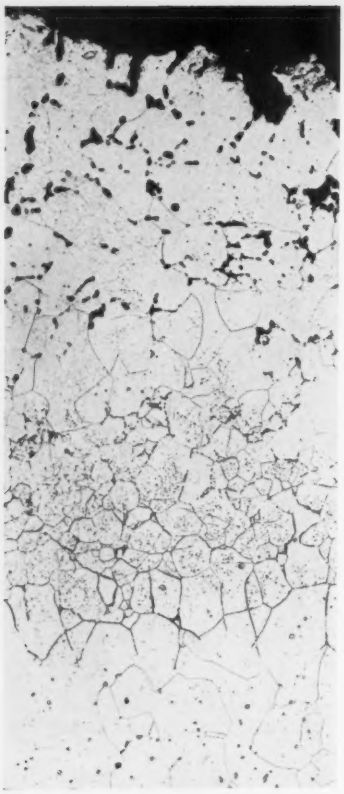

Fig. 9—Peel on inner wall of pinhole of track link annealed in ore containing 0·5% of sulphur, with graphite inside pinhole × 60

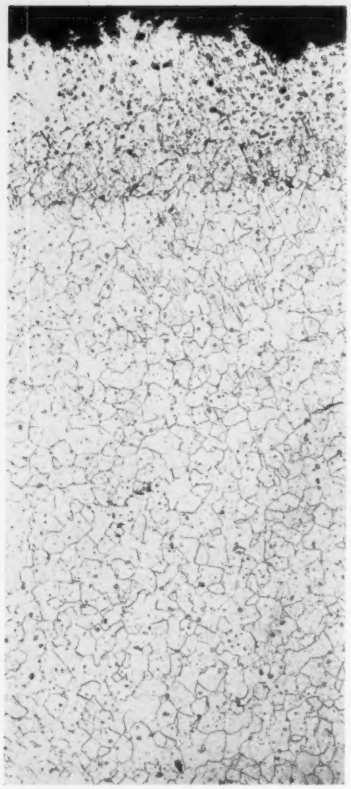

Fig. 10—Outer surface of pinhole shown in Fig. 9 × 60

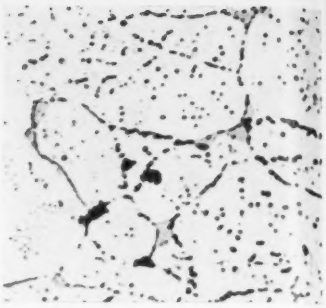

Fig. 11—Peel on unhardened track link ; section near inner surface, showing iron sulphide × 500

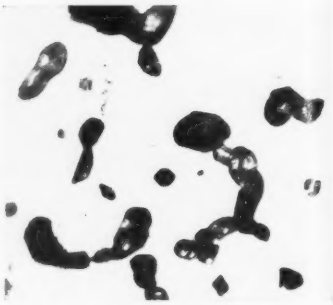

Fig. 12—Peel on unhardened track link, showing fayalite × 500

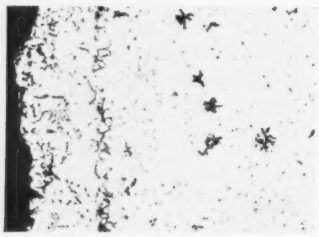

Fig. 13 × 80

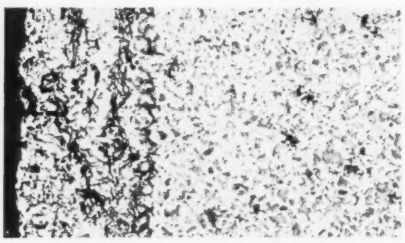

Fig. 15 × 80

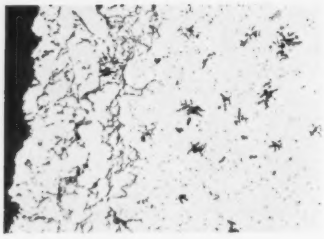

Fig. 14 × 80

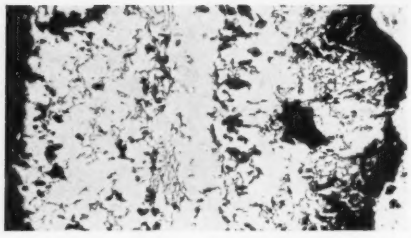

Fig. 16 × 80

Figs. 13 to 16—Progressive stages in the development of peel (reduced to one-half linear in reproduction)

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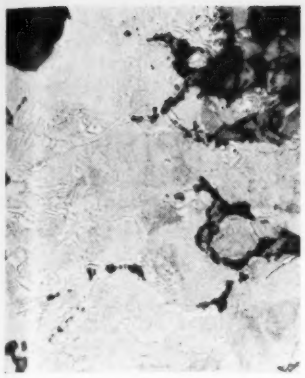

Fig. 17 × 500

Fig. 18 × 500

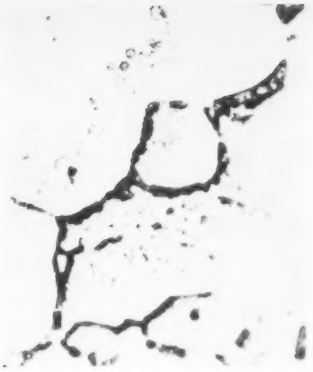

Fig. 19 × 500

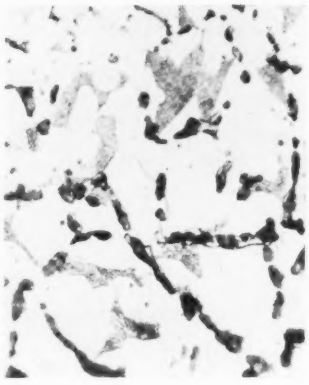

Fig. 20 × 500

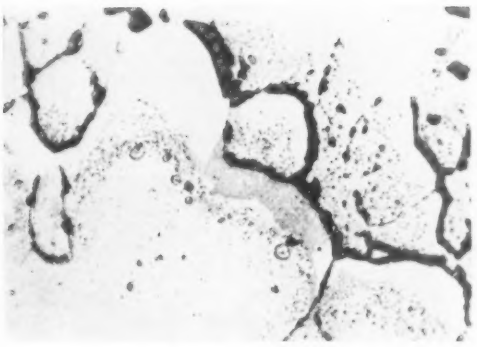

Fig. 21 × 500

Figs. 17 to 21—Typical structures in peeled areas, showing silicates, sulphides, and carbides

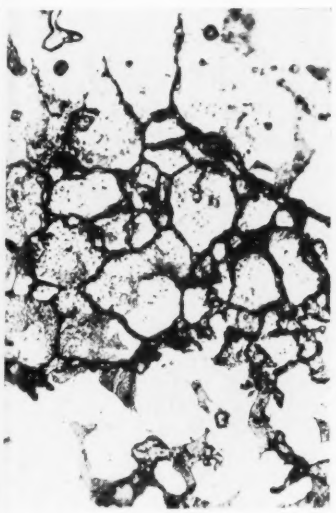

Fig. 22 × 500

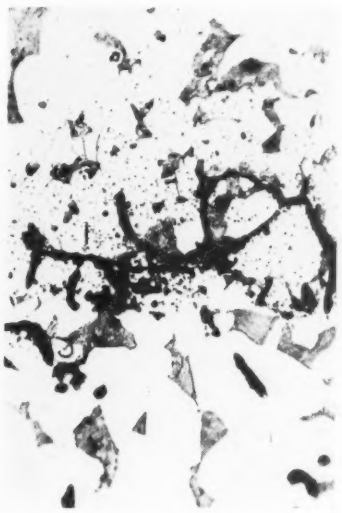

Fig. 23 × 500

Fig. 24 × 2500

Figs. 22 to 24—Typical structures near the peel/metal interface
(Micrographs reduced to three-quarters linear in reproduction)

[*Bowden*

Fig. 25—Silicon-manganese steel (Si 1·8%, Mn 0·8%, S 0·04%), after annealing for 120 hr. in 80% of spent ore plus 20% of coke. (a) Annealed in hot part of oven; specimen very brittle. (b) Peeled; specimen tough, and bent through 120° before fracture

Fig. 26—Edge structure of brittle silicon — manganese steel specimen shown in Fig. 25 (a) × 60

Fig. 27—Edge structure of peeled silicon — manganese steel specimen shown in Fig. 25 (b) × 60

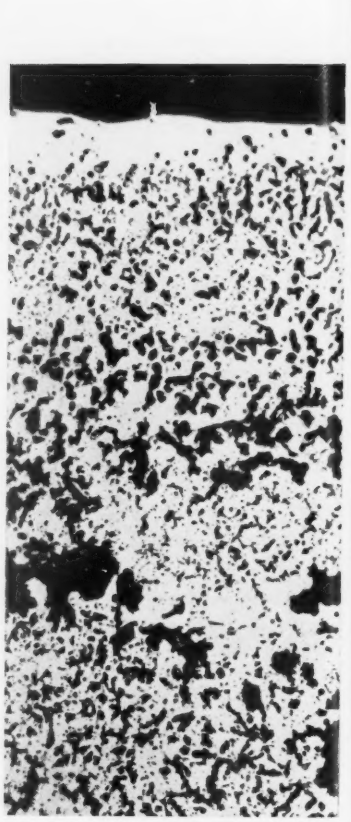

Fig. 28—Edge structure of grey cast iron after "malleablizing" (Si 2·1%, Mn 0·39%, S 0·12%, P 0·56%). Note large amount of oxidation products. Unetched × 60

In another works peel was detected in cruciform grids shortly after the change-over from track links to that type of casting. It was discovered that the strength of the ore mixture had been reduced from 4 : 1 (spent ore to new ore) to 20 : 1, in order to compensate for the thinner sections of the smaller castings, and that furnaces had been damped down during the later stages of the malleablizing process in an endeavour to economize in fuel and prolong the life of the annealing pots. The ore mixture was restored to the old ratio, the flues were opened to promote a more buoyant atmosphere, and the peeling troubles disappeared. Previous attempts to produce peel in these ovens during the works experiments referred to had been unsuccessful, except in those cases where localized concentrations of carbon monoxide had been artificially created.

The Managing Director of a works where peeling troubles had been particularly prevalent reported recently that the firm have been completely free from peel since strengthening the ore mixtures and opening the flues over $2\frac{1}{2}$ years ago.

The composition of the gas phase inside annealing pans has been repeatedly studied, and in normal practice the main components are carbon monoxide and carbon dioxide, in a ratio of about 70 : 30, together with small quantities of hydrogen, nitrogen, and methane.[3]

THEORY OF PEEL FORMATION

As a result of the works experiments described above, a theory was advanced suggesting that peel formation, which takes place during the later stages of the malleablizing process, occurs when the annealing atmosphere contains abnormally high proportions of carbon monoxide and that peel is formed by the preferential oxidation of the silicon in the decarburized metal to fayalite ($2FeO.SiO_2$) by carbon monoxide (CO) and carbonyl sulphide (COS). The reaction may proceed according to the following equations:

$$4CO + Si + 14Fe \rightarrow 2FeO.SiO_2 + 4Fe_3C \quad \ldots\ldots(1)$$

$$CO + FeS \rightarrow COS + Fe \quad \ldots\ldots(2)$$

$$4COS + Si + 18Fe \rightarrow 2FeO.SiO_2 + 4FeS + 4Fe_3C \quad \ldots\ldots(3)$$

Microscopical examination revealed the presence of iron sulphide, as in Fig. 11, and of considerable numbers of coalesced particles (yellow to olive-green or black) which had the appearance of fayalite, as in Fig. 12. Examination at the National Physical Laboratory confirmed the presence of fayalite, ferrous sulphide, and an iron sulphide in the unusual form of ferrous sulphide with a superlattice structure ($FeS + xS$).

The reactions quoted, which were assumed to move progressively inwards as the silicon becomes oxidized from the surface layers, would explain the presence of Fe_3C and FeS at the peel/metal interface (*see* Figs. 17 to 24). Equation (2) was assumed to represent the reaction between carbon monoxide in the annealing atmosphere and sulphur in the annealing ore or in the surface layers of the casting. Equations (2) and (3), which would not necessitate a sulphur " pick-up " by the casting, would explain the concentration of the original sulphur from the outer layers of the casting and the film of silicates at the peel/metal interface. Under certain conditions peel formation may be accompanied by migration of the sulphur from the annealing ore or the furnace gases to the castings (*see* Fig. 1 (*e*) and (*f*)), but chemical analyses made by Ingall and Field show that peeling does not necessarily involve an increase in the sulphur content of the metal. Jones found that the sulphur content of a specimen of peel was variable but gave lower values than the core metal. Similar sulphur-rich layers, observed by Preece[4] at the scale/metal interface of a 4% silicon steel, may have been formed by similar preferential oxidation reactions.

Micro-examination suggests that silicates are first formed as minute discrete particles throughout the iron grains (*see* Figs. 13 to 24). As grain growth occurs the silicate particles are swept forward by the boundaries of the expanding grains and coalesce, until they finally form the more massive particles seen at the grain boundaries. This coalescence of particles which are insoluble in the metal is presumably due to the presence of a molten eutectic, in which the silicates are soluble, at the grain boundaries.

It is known that iron sulphide (melting point 1200° C.) forms eutectics with (*a*) iron (69% of iron and 31% of sulphur, melting point 985° C.), (*b*) iron oxide (42% of FeO and 58% of FeS, melting point 940° C.), and (*c*) iron silicate (melting point 990° C.), and it is considered probable that in such a complex system as cast iron other eutectics having even lower melting temperatures may be formed. Such eutectics would be mobile at high malleablizing temperatures.

During cooling from the annealing temperatures recrystallization occurs ($\alpha \rightarrow \gamma$ iron) and new grain boundaries are formed. These boundaries do not usually coincide with those at which the silicate particles collected and coalesced.

Progressive stages in the development of peel formation are illustrated in Figs. 13 to 16. Typical structures observed in peeled specimens are shown in Figs. 17 to 24 and illustrate the association of pearlite with the silicate and sulphide inclusions.

When a piece of $\frac{7}{16}$-in. dia. silicon–manganese steel rod of specification STA 2(*b*) was packed in a mixture of 80% of spent annealing ore (sulphur 0·2%) and 20% of coke, and heated through a malleablizing cycle, a definite layer of peel was produced. The steel rod was tough and bent through 120° without fracture ; this rod is shown in Figs. 25 (*b*) and 27. Another piece, cut from the same bar and heated under similar conditions in a hotter part of the same oven, was not peeled but developed a very brittle structure (*see* Figs. 25 (*a*) and 26). The freedom from peel may be due to the fact that this piece of steel was in the top pot of a stack which was exposed to the oxidizing influence of furnace gases. Peel was also produced on silicon–manganese steel by Preece, who heated a specimen at 1000° C. for 50 hr. in an atmosphere of carbon monoxide plus 0·3% of sulphur.

These experiments suggest that the manganese/sulphur ratio in the original metal does not affect the rate of peel formation, and that the retarding effect of manganese reported by other investigators may be due to the fact that the sulphides and silicates of manganese melt at higher temperatures than the corresponding compounds of iron, and that manganese and iron sulphides form a eutectic which, unlike the

iron/iron-sulphide eutectic, would be infusible at malleablizing temperatures.*

Analysis of results obtained by Timmins supports the view that the rate of peel formation increases as the silicon/manganese ratio of the metal increases. Since silicon and manganese are oxidized in peel formation, it is consistent that the rate of reaction should depend upon the composition of the metal and the extent to which the silicates and silicate complexes produced are fusible at malleablizing temperatures.

Considerable " growth " and internal oxidation, which were observed when a piece of grey cast iron was " malleablized " in a normal ore mixture (*see* Fig. 28) may have been due to preferential oxidation of silicon by carbon monoxide according to equation (1).

It is known that the amount of " growth " which occurs when grey iron castings are heated in oxidizing atmospheres increases with increased silicon content. This growth may be attributed to the formation of silicates due to the preferential oxidation of silicon, within the casting, by carbon monoxide produced by reaction between graphite and the oxidizing gases which diffuse into the metal.

SUMMARY, CONCLUSIONS, AND RECOMMENDATIONS

Peeling of whiteheart malleable-iron castings occurs when the malleablizing atmospheres contain abnormally high proportions of carbon monoxide. These conditions may be avoided by the use of strong annealing ore mixtures and the maintenance of a mildly oxidizing furnace atmosphere. A smoky flame should be avoided and reducing conditions should not occur at any time during the anneal.

The following recommendations concerning metal composition are considered desirable :

(1) The sulphur content should be as low as possible.

(2) The manganese content should be relatively high, a figure of 0·4% being suggested as an upper limit.

(3) The silicon content should be kept as low as possible, consistent with ready response to malleablizing.

It is considered that the epidemic of peeling troubles which was in certain quarters attributed to the use of Riff ores was primarily due to the use of weak ore mixtures and insufficiently oxidizing atmospheres in the malleablizing ovens.

At certain stages of the war there was an acute shortage of hematite ores and the supply position threatened to become desperate. Ore mixtures, which had previously been made up on a 4 : 1 or 6 : 1 basis, were reduced in strength, and additions of new hematite ore were in many cases restricted to the amounts necessary to make up losses in bulk. Some plants found it necessary to add crushed slag, gravel, and other inert media in order to obtain the bulk of packing material. The old practice of " weathering " spent ores for prolonged periods was abandoned through shortage of ore, and the spent ores went back into circulation without being re-oxidized. Many firms attempted to economize in fuel and to increase pot life by closing-down flues and maintaining a sluggish smoky flame.

All these factors contributed to a steady reduction in the strength of annealing ore mixtures and helped to build up conditions favourable to the formation of peel.

During the war, emergency relaxations permitting peel to a depth not exceeding 0·016 in. on carrier-track links were granted, in order to obtain the necessary supplies of castings. Representatives of the industry agree that peeling is an avoidable defect, and it is considered that this concession should now be revoked.

Acknowledgments

Acknowledgments are due to Dr. A. L. Norbury, who supervised this work, to Mr. J. H. Geeves, who was closely associated with all the experiments, and to Messrs. John Maddock and Co., Ltd., the Walsall Conduits, Ltd., the Lilleshall Co., Ltd., Messrs. J. Birch and Sons, Ltd., and the Halesowen Brick and Tile Co., Ltd., for the provision of materials and facilities. Acknowledgments are also made to the Chief Scientist, Ministry of Supply, for permission to publish this paper.

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* The melting point of manganese sulphide is 1610° C. ; the manganese-sulphide/ferrous-sulphide eutectic (composition, 6·5% of MnS + 93·5% of FeS) melts at 1164° C.

PEELING OF WHITEHEART MALLEABLE IRON

By G. R. Webster, A.M.I.Mech.E., A.M.I.Mar.E., F.I.M., and A. E. Probst

SYNOPSIS

Peeling, as distinct from scaling, is described, and examples of the defect are given and discussed. Experiments with malleable iron of varying thicknesses and with various sulphur contents have shown that a high sulphur content in the iron ore used for malleablizing causes peeling.

Introduction

THE phenomenon of peeling of whiteheart malleable-iron castings has undoubtedly been well known throughout the industry for many years, and in the works with which the authors are connected it has occurred to a more or less extent for brief periods on infrequent occasions, disappearing again almost as quickly and mysteriously as it arose. It was observed, however, that on these occasions the sulphur content of the metal was generally higher than normal, and this was believed to be one of the reasons for the defect. The presence of sulphide inclusions in the boundary of the peeled zone and the sound part of the casting, seen in numerous microscopical examinations, seemed to substantiate this opinion. A high percentage of dolomite in the annealing ore was believed to be another cause, though in this case the nature of the peel was quite different and was of a localized character, and its presence could be observed on the face of the casting without having to distort it in any way.

During the recent war this "peeling disease" suddenly reached somewhat alarming proportions and did not disappear after a short period, as it had on previous occasions. Again no particular reason for its re-occurrence was obvious, as the foundry and malleablizing practices had remained constant. The analysis of the cast product was normal. It was only then that the annealing ore came under suspicion, though nothing could be established immediately; a reason for the suspicion was the fact that some months earlier the supplies of Cumberland ore for the malleable-iron industry ceased completely, foreign ores being substituted. Initially, Thabazambi ore from South Africa was supplied, and during this period no peeling occurred. However, after the invasion of North Africa by the allies, Riff ore became available, and was then the only type of ore supplied to the industry. It was during this period that peeling became extensive.

The investigation described in this paper consisted of practical works' experiments carried out during the war at a vital period of track-link production when the incidence of peeling was abnormal. The experiments are recorded in chronological order, and they were not intended as part of a planned long-term research programme.

Description of Peel

The very serious peeling here referred to must be distinguished from the general scaling which involves the whole surface of the casting and which is not necessarily connected with the sulphur content of the malleable iron, and which can occur when a practically sulphur-free iron ore is used in the malleablizing process. This scaling is chiefly caused by incorrect composition of the malleable iron, *e.g.*, an abnormal sulphur/manganese ratio, particularly when this is associated with an unusually high silicon content, and it becomes more apparent when a concentrated ore mixture is used.

The particular type of peel here discussed is not spread over the whole of the surface of the casting but is of a localized character, and can be recognized by small and large metallic "blisters" in lentil-like formation on the surfaces of the castings after their removal from the malleablizing-furnace pots. If these castings are deformed, the lentil-like particles—which may consist of one or more separate layers and which are low in ductility and strength—break away from the surface, and slight, fairly smooth, surface cavities of a grey oxidized appearance remain. In some peeled castings, however, the surface appearance is normal, and the defect is apparent only after mechanical deformation. If such faults occur in isolated positions on large thick-walled castings they can be classified as reducing only the appearance of the piece and not necessarily its engineering usefulness in the final product, whereas with thin-walled castings the peel may form a considerable portion of the total wall thickness, as was the case with *TL*2948-type Bren-gun-carrier links, where peeled castings had to be regarded as scrap. In cases where castings undergo a further surface treatment, such as hot-galvanizing, the blisters become detached in the acid pickling bath from the base metal and the resultant galvanized piece shows a corrosion surface-pitted appearance.

EXPERIMENTAL WORK

All the castings and test-pieces used in the experiments to be described were malleablized in a town's-gas-fired continuous bogie-type kiln giving a very even temperature and constant atmosphere. The pots, packed with castings and ore, were sealed and stacked on the bogie hearth. The castings were all packed in the same relative position throughout the experiments, and the malleablizing cycle was as follows:

Preheating to 980° C.	34 hr.
Soaking at 980° C.	70 hr.
Cooling from 980° to 800° C.	40 hr.
Total time	144 hr.

Manuscript received 21st April, 1947.

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All the pots were withdrawn from the kiln at a temperature of approximately 800° C.

The new and spent ores referred to in the accounts of the experiments were of the Riff type. Representative analyses were as follows:

	New Ore	Spent Ore
Fe, %	0·73	29·28
FeO, %	2·01	54·75
Fe_2O_3, %	84·94	1·10
SiO_2, %	6·07	12·98
CaO, %	1·23	0·88
MgO, %	0·74	0·62
Sulphur, % ...	0·12	0·083

The sulphur content of the new Riff ore varied considerably; analyses of samples taken from various trucks showed sulphur contents of 0·21, 0·189, 0·202, and 0·140%.

Preliminary Experiments

Preliminary experiments were carried out to ascertain the effects of high and low sulphur contents in the metal, strengths of ore mixtures, segregation of new ore from spent ore so that the castings were surrounded only by spent ore, etc. In all these experiments peeling still occurred to practically the same degree as in the normal malleablizing practice.

Experiment 1

Although the preliminary experiments had not been conclusive regarding the causes of peeling, some of them had indicated that impurities present in the ore must be responsible for the phenomenon. On a closer examination of the new ore it was found that there was a considerable variety of suspicious pieces present, and certainly some which were contaminated with pyrites (approximately 5% of sulphur), as shown in Fig. 1. Other pieces had a characteristic white sparkling surface, and still others were found to be of low specific gravity and had the general appearance of coke. For the purposes of experiment, some of these substances were hand-picked from the deliveries of new ore.

In order to ascertain the influence of these various impurities in the ore upon the cause of peeling, some 70 castings of circular cross-section, $1\frac{1}{4}$ in. in outside dia. and with a wall thickness of $\frac{5}{32}$ in., were packed in spent ore in a separate steel box and placed in the usual malleablizing pot, together with the normal production castings, the usual ore mixture of 1 : 10 being employed. All these experimental castings were of the same analysis, *viz.*:

Carbon	3·20%
Silicon	0·74%
Manganese	0·25%
Phosphorus	0·058%
Sulphur	0·186%

The 70 experimental castings were split into seven batches of ten, grouped together, and placed side by side in the steel box. On top, and touching each casting, were placed several approximately $\frac{1}{4}$-in. pieces of the following substances: (1) New ore contaminated with pyrites (5·5% of sulphur), (2) new ore with the white sparkling surface, (3) new ore with the porous surface, (4) limestone, (5) gas coke, (6) ferrosulphur (27·7% of sulphur), and (7) spent ore only. The results of this experiment are shown in Table I, and suggested that the sulphur content of the ore was responsible for the peeling.

From Table I it will be seen that the castings in batch 6, in which two or three small pieces of ferrosulphur were placed on the top of each casting, actually peeled all over, even on the bottoms and in the insides of the castings, where no ferrosulphur was present. It would appear that the sulphurous gas produced from the ferrosulphur attacked the metal wherever it came into contact with it. It is also more than likely that the heavy peeling observed in four of the castings of batch 5 was produced by the sulphurous gas from the ferrosulphur in batch 6, as batch 6 was adjacent to batch 5 in the box. In other words, the sulphurous gases from batch 6 have also attacked some of the castings in batch 5. Batch 1, to which new ore contaminated with pyrites was added, produced the same type of peeling which had been found to be giving trouble in the normal production. The castings in this batch had not peeled all over, but only on the tops and sides and particularly around the pyritic ore, whereas the bottoms in most cases showed no peeling whatsoever. The extent of the peeling in batches 1 and 6 is illustrated in Fig. 2. The peel on the castings in batches 2, 3, 4, 5, and 7 was only very localized, and it is extremely doubtful whether it was produced by any of the substances with which these batches had been packed. From examination of batch 7, however, it appeared that even the spent ore seemed likely to produce peeling.

Experiment 2

In order to ascertain the effect of pyritic ore on peeling after one malleablizing cycle, it was decided to produce a certain amount of this type of ore. Three castings (not malleablized) were therefore assembled to make a canister, the bottom half of which was filled with new pyritic ore; the canister was then

Table I

EXPERIMENTAL DATA FOR TEST-CASTINGS MALLEABLIZED IN CONTACT WITH VARIOUS SUBSTANCES

Batch No.	Substance	Results
1	New ore contaminated with pyrites	Top half of each casting peeled approx. $\frac{1}{8}$ in. All 10 pieces affected
2	New ore with white sparkling surface	No peeling. All 10 pieces good
3	New ore with porous surface	7 pieces good; 3 pieces with localized peeling
4	Limestone	All good
5	Gas coke	3 pieces good, 3 slightly peeled, and 4 with heavy peeling
6	Ferrosulphur	All pieces very badly peeled over entire surfaces. Thickness of peeling in some areas equal to the thickness of the castings
7	Spent ore	6 pieces good and 4 with localized peeling

almost completely sealed and placed in a malleablizing pot, which was otherwise packed in the usual manner (ore ratio 1 : 10). One malleablizing treatment under these conditions reduced the sulphur content of the new pyritic ore from 5·5% to 3% (spent ore).

As was expected, this canister peeled rather heavily, as shown in Fig. 3, owing to the very high concentration of sulphurous gases from the pyritic ore. The peeling was concentrated in the joints of the three assembled parts of the canister, where the produced gases escaped, and particularly in the top portion of the canister, which was not actually in contact with the pyritic ore.

Experiment 3

Experiments 1 and 2 very clearly showed that new pyritic ore could cause peeling, and experiment 2 in particular showed that peeling can be produced even when the pyritic ore is not in direct contact with the castings. In order to verify the results of these first two experiments and to investigate the effect of spent pyritic ore on peeling, 40 castings of the same type as used in experiment 1, but of different composition, were used in further tests. The analysis of these castings was as follows :

Carbon	3·44%
Silicon	0·67%
Manganese	0·21%
Phosphorus	0·057%
Sulphur	0·167%

The 40 castings were separated into four batches of ten and placed in a box, as in experiment 1, in spent ore. The procedures were as follows :

Batch No.

8 A few pieces of new pyritic ore (sulphur 5·5%) were placed about half an inch above the castings.

9 Similar to batch 8, but using spent pyritic ore (sulphur 3%).

10 Several pieces of new pyritic ore (sulphur 5·5%) were placed touching the top surface of each casting.

11 Similar to batch 10, but using spent pyritic ore (sulphur 3%).

After treatment, all the castings showed peeling to a greater or lesser degree on the surface where the new and spent pyritic ore had been placed. The extent of the peeling was larger in batches 10 and 11, where the pyritic ore was in actual contact with the casting. In batches 8 and 9, where the new and spent pyritic ore was placed half an inch above the castings, peeling was of a localized character, although the thickness of the peel was not less than in batches 10 and 11. The extent of the peeling in the batches mentioned is shown in Figs. 4 and 5. This experiment proved that it is not necessary for high-sulphur ore to be in actual contact with the castings for peeling to occur.

Table II

COMPOSITION OF STEPPED TEST-BARS

Cast	Carbon, %	Silicon, %	Manganese, %	Phosphorus %	Sulphur, %
A	3·23	0·73	0·22	0·054	0·165
B	3·23	0·73	0·22	0·054	0·193
C	3·23	0·73	0·22	0·054	0·241
D	3·23	0·73	0·22	0·054	0·281
E	3·25	0·73	0·22	0·054	0·346
F	3·25	0·73	0·22	0·054	0·374

Experiment 4

It was next decided to investigate the effect of the new pyritic ore on malleable iron of various thicknesses and sulphur contents. For this purpose a number of stepped test-bars of the type shown in Fig. 6 were cast from malleable iron of different sulphur contents, as shown in Table II. The stepped bars were 10·5 in. long and 1 in. in width, and the steps varied in thickness from $\frac{1}{8}$ in. at one end to $\frac{3}{4}$ in. at the other, giving six different thicknesses, namely, $\frac{1}{8}$, $\frac{1}{4}$, $\frac{3}{8}$, $\frac{1}{2}$, $\frac{5}{8}$, and $\frac{3}{4}$ in. The length of each step was $1\frac{3}{4}$ in.

Two test-bars of each composition were packed in spent ore in a separate steel box, as before. On the top of one test-bar from each cast a few pieces of pyritic new ore (sulphur approximately 5·5%) were placed on the centre of each step. The duplicate bar of each composition was packed in spent ore only. The whole box was placed in a malleablizing pot as before (ore ratio 1 : 10). The analysis of the spent ore used was :

Fe	29·28%
FeO	54·75%
Fe_2O_3	1·10%
SiO_2	12·98%
CaO	0·88%
MgO	0·62%
Sulphur	0·083%

After malleablizing, each test-bar was cross-sectioned through the centre of each step, and each specimen thus obtained was then polished in the usual way and etched with 20% nitric acid. The results obtained were as follows :

(1) No peeling whatsoever could be detected in any of the test-bars which were malleablized in the spent ore.

(2) All test-bars which were malleablized with pyritic new ore showed peeling to a larger or smaller extent, although the surfaces of all these test-bars showed no actual sign of peeling.

(3) The thickness of the peel was heavier in the small $\frac{1}{8}$-in. section than in any of the other sections.

(4) There was a slight tendency for the peel to decrease with the increase in the section from $\frac{1}{4}$ to $\frac{3}{4}$ in.

(5) The sulphur content of the malleable iron appeared to have very little influence on the peel, particularly in the 0·165–0·281% sulphur range, although test-bars from casts *E* and *F*, with abnormally high sulphur contents, appeared to be slightly less peeled than the other test-bars.

The extent of peeling on the $\frac{1}{8}$, $\frac{1}{4}$, and $\frac{3}{8}$-in. sections with various sulphur contents is shown in Figs. 7, 8, and 9, respectively. An indication of the thickness of the peel on the various test-bars, measured on the face where the new pyritic ore had been positioned is given in Table III. The figures given indicate the average thickness, in millimetres, of the peel on stepped test-bars malleablized in spent ore plus a few pieces of new pyritic ore (sulphur 5·5%) in

Table III

THICKNESS OF PEEL (mm.) ON STEPPED TEST-BARS MALLEABLIZED IN SPENT ORE

Cast	Sulphur, %	Test-Bar Section, in.					
		⅛	¼	⅜	½	⅝	¾
A	0·165	1·0	0·6	0·5	0·6	0·5	0·5
B	0·193	1·0	0·7	0·6	0·55	0·5	0·6
C	0·241	1·1	0·7	0·6	0·65	0·5	0·5
D	0·281	1·0	0·5	0·5	0·5	0·4	0·5
E	0·346	0·8	0·4	0·4	0·35	0·4	0·3
F	0·374	0·75	0·45	0·4	0·25	0·3	0·3

actual contact with the test-bars. It must be stressed that the thicknesses given represent only approximate averages, and they cannot be taken as being absolutely accurate.

Experiment 5

In experiment 4, very weak spent ore was used, which obviously could not have produced very oxidizing conditions. It was therefore considered desirable to repeat this experiment with a stronger ore mixture.

Another box of stepped test-bars was packed in the same way as in experiment 4, but using an ore mixture of one part of new ore to five parts of spent ore. However, the ore actually surrounding the test-bars was spent ore only; this was thought necessary to ensure that no new ore of high sulphur content could come into contact with the test-bars. A few small pieces of new pyritic ore were placed on the top of each step, as in experiment 4.

After treatment, all the test-bars showed slight whitish surface oxidation, particularly on the ⅛- and ¼-in. sections and along the edges of the specimens. The thickness of the peel was found to be slightly larger than in experiment 4, particularly on the thin sections, as shown in Table IV and Figs. 10 and 11. The microstructure of a ¼-in. section from cast *C* is shown in Figs. 12 and 13.

Experiment 6

The next experiment was carried out to determine the loss of sulphur by repeated malleablizing treatments of new pyritic ore (total sulphur approximately 5%) hand-picked from the bulk supply of normal Riff ore. These ore samples were packed in a 6-in. length of 1½-in. gas tubing, which was then sealed at both ends. A series of small holes was drilled through the wall of the tubing, around the periphery, and the tube and its contents were placed in a normal malleablizing pot. These arrangements permitted the atmosphere from the pot to percolate into the tube and come into direct contact with the ore samples.

The various malleablizing treatments under these conditions are shown in Table V and Fig. 14, and it will be seen that the sulphur content of the new pyritic ore was reduced in each stage of the malleablizing cycle. After the first malleablizing treatment the loss of sulphur was considerable, as would be expected.

Table IV

THICKNESS OF PEEL (mm.) ON STEPPED TEST-BARS MALLEABLIZED IN SPENT-ORE/NEW-ORE MIXTURE

Cast	Sulphur, %	Test-Bar Section, in.					
		⅛	¼	⅜	½	⅝	¾
A	0·165	1·1	0·7	0·7	0·4	0·5	0·5
B	0·193	1·1	0·7	0·7	0·5	0·5	0·5
C	0·241	1·3	0·8	0·6	0·5	0·5	0·6
D	0·281	1·3	0·5	0·5	0·5	0·5	0·5
E	0·346	1·3	0·5	0·5	0·3	0·3	0·3
F	0·374	1·3	0·4	0·3	0·3	0·3	0·3

Table V

LOSS OF SULPHUR DURING REPEATED MALLEABLIZING TREATMENTS

Number of Times Malleablized	Fixed Sulphur		Volatile Sulphur		Total Sulphur	
	Content, %	Loss	Content, %	Loss	Content, %	Loss
New	1·30	...	3·83	...	5·13	...
1	0·16	1·14	1·64	2·19	1·80	3·33
2	0·084	0·076	1·53	0·11	1·614	0·186
3	0·068	0·016	1·48	0·05	1·548	0·066
4	0·065	0·003	1·42	0·06	1·485	0·063

The subsequent treatments showed a very small loss in both the fixed and volatile sulphur contents.

The higher losses in sulphur in the first malleablizing treatment as compared with the sulphur loss experienced in experiment 2 can be explained by the fact that in experiment 2 the test-canister was almost completely sealed and therefore the ore samples were not exposed to the normal atmosphere of the malleablizing pot, whereas in the present experiment the gases in the malleablizing pot had free access to the pyritic ore in the perforated tube.

SUMMARY OF RESULTS

The results of the experiments may be summarized as follows:

(1) The sulphur content of the new and of the spent ore has a very great effect on peel formation.

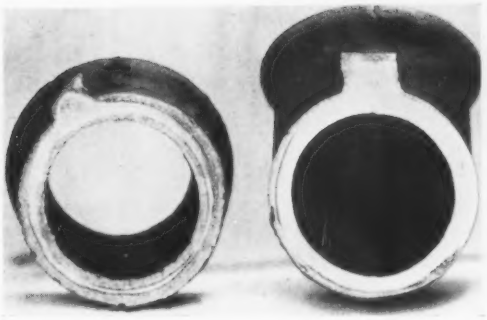
(2) The sulphur contents of the particular pieces of ore used were of approximately 5% and 3% respectively, both of which produced peeling to the same extent.

(3) High-sulphur ore produces peeling even if the ore is not in actual contact with the castings.

(4) The extent of peeling is slightly higher when using a strong ore mixture, although differences in the thickness of the peel as produced by a weak and strong ore mixture are not as pronounced as would be expected.

(5) Thin sections of the order of ⅛ in. are apt

Fig. 1—New Riff ore contaminated with pyrites



Batch 6 Batch 1

Fig. 2—Samples from batches 6 and 1

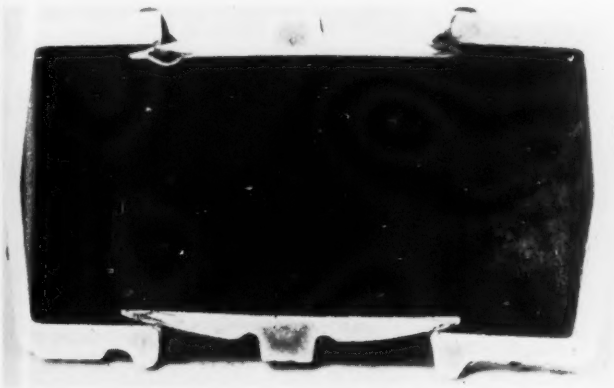
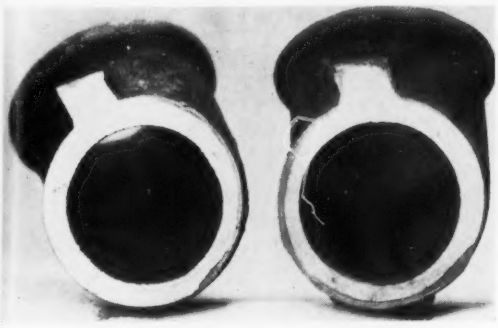


Fig. 3—Peeling in experimental canister

Batch 8 Batch 9

Fig. 4—Samples from batches 8 and 9



Batch 10 Batch 11

Fig. 5—Samples from batches 10 and 11

Fig. 6—Stepped test-bar

(Figs. 2 to 5 etched with 20% nitric acid)

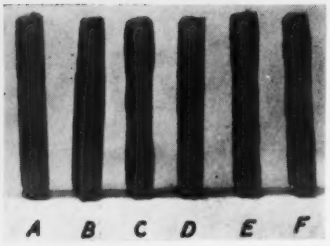


Fig. 7—Experiment 4 ; ½-in. section test-bars

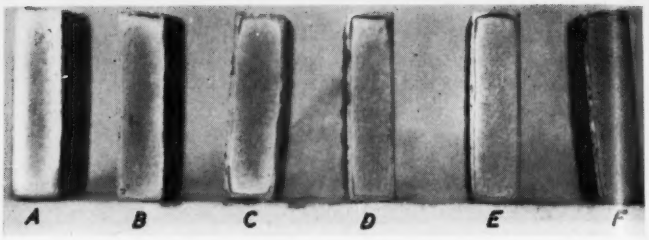


Fig. 8—Experiment 4 ; ¼-in. section test-bars

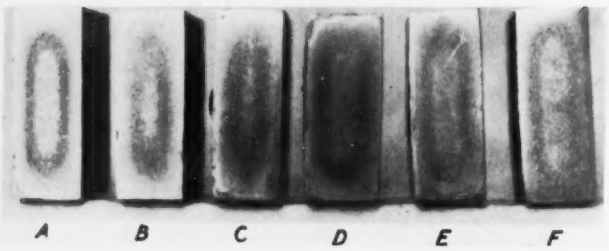


Fig. 9—Experiment 4 ; ⅜-in. section test-bars

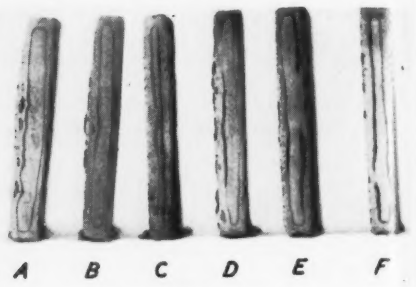


Fig. 10—Experiment 5 ; ⅛-in. section test-bar

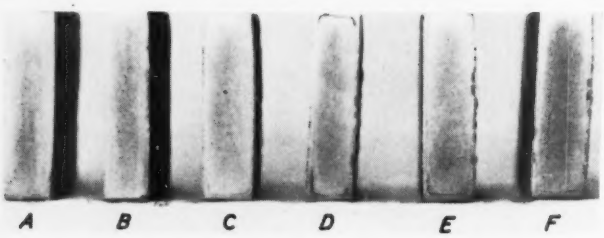


Fig. 11—Experiment 5 ; ⅛-in. section test-bars

Figs. 7 to 11—Test-bars used in experiments 4 and 5 ; sulphur contents (%) : B 0·193 ; C 0·241 ; D 0·281 ; E 0·346 ; F 0·374

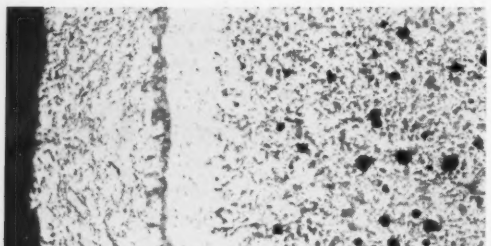


Fig. 12 × 600

Fig. 13 × 600

Figs. 12 and 13—Microstructure of ¼-in. section test-bar from cast C

to peel more than heavier sections. There definitely appeared to be a certain relation between the section of the casting and the thickness of the peel.

(6) The sulphur content of the metal did not seem to influence the thickness of the peel to any noticeable degree, although metal with a sulphur content of more than 0·28% gave a thinner peel, particularly on heavier sections.

(7) The sulphur content of the Riff ore was very inconsistent. Although the average analysis of a truck varied between 0·12% and 0·2%, the sulphur content of some hand-picked pieces of ore was as high as 5%.

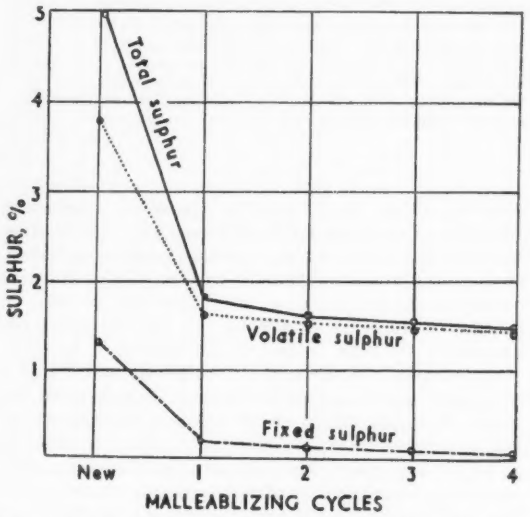


Fig. 14—Loss in sulphur content of pyritic Riff ore after several malleablizing cycles

CONCLUSIONS AND DISCUSSION

No definite explanation can be given as to the mechanism of peeling as produced in these experiments. It is known that iron pyrites (FeS_2) gives up sulphur when heated in an oxidizing atmosphere above a certain temperature. It also appeared that the ferrosulphur as used in experiment 1 was oxidized by the pot atmosphere, probably to SO_2. However, SO_2 is a reducing agent, and one would therefore not expect that it would attack the metal during the malleablizing process.

A microscopical examination of peeled test-bars revealed a large number of oxide inclusions in the peel, whereas in the boundary of the peel and the parent metal a fairly large number of sulphides was found. The sulphur contents of the peeled and sound metal of batch 1 (experiment 1) were also determined, and 0·168% of sulphur was found in the peel and 0·387% in the parent metal, whereas the metal as cast had 0·186% of sulphur. This does indicate that peeling is associated with a segregation of sulphur to the parent metal underneath the peel, which is a well-known phenomenon in the whiteheart malleable-iron industry.

The possible explanation of the mechanism of peeling may therefore be that some reactions take place when the SO_2 from the ore comes into contact with the surface of the casting at the higher malleablizing temperatures, thus :

$$3Fe + SO_2 \longrightarrow FeS + 2FeO$$
$$4Fe + Fe_3C + 3SO_2 \longrightarrow 3FeS + 4FeO + CO_2$$

Naturally these reactions are of a complex nature and must be considered in conjunction with the decarburization of the castings by the atmosphere in the malleablizing pot. Therefore the above reactions are tentatively put forward as being possible, but by no means certain.

It has also been found by close observation of two malleablizing kilns that the extent of peeling (percentages of peeled castings) bears a certain relationship to the sulphur content of the spent ore. In No. 1 malleablizing kiln, where the ore ratio used was 1 : 10 and peeling occurred to a very small degree, the sulphur content of the spent ore was found to be between 0·057 and 0·083%. These analyses were spread over a period of several months. In No. 2 malleablizing kiln, where the ore ratio was gradually reduced from 1 : 4 to 1 : 7, the sulphur content of the spent ore was found to decline from 0·129 to 0·100%. The extent of the peeling was heaviest when the sulphur content of the spent ore was of the order of 0·129% and it gradually lessened as the sulphur content declined to approximately 0·1%, although the peeling was still heavier than in No. 1 malleablizing kiln, another reason being a longer soaking time at 980° C. The sulphur content of the spent ore could be reduced in both kilns by substituting up to 25% of the spent ore with granite chippings, and consequently the extent of peeling was quite considerably reduced but not entirely prevented.

In order to have further proof of the effect of the sulphur content of the new annealing ore on peeling, another experiment was carried out at the time with Cumberland hematite ore containing 0·009% of sulphur. Approximately one ton of the Cumberland ore was first malleablized, in order to reduce its Fe_2O_3 content. It was then used for the malleablizing of castings of the normal production with an ore ratio of 1 : 10. After several malleablizing treatments using this Cumberland ore with a normal addition of new ore after each cycle and with careful inspection of the castings in conjunction with physical tests, no sign of peeling was observed.

It is beyond doubt that the high sulphur content of the Riff ore was the main cause of peeling. It can also be stated that peeling would not have disappeared so long as this ore had to be used in malleablizing, although the extent of the peeling could be reduced by diluting the spent ore with granite chippings.

Acknowledgments

Acknowledgments are due to the Management of the Britannia Iron and Steel Works, Ltd., Bedford, for permission to publish this paper, and also to the British Cast Iron Research Association for the two micrographs reproduced as Figs. 12 and 13.

EXPERIMENTAL WORK CARRIED OUT AT THE BRITISH CAST IRON RESEARCH ASSOCIATION

By J. Bernstein, A.R.I.C., A.C.T.C.(Birm.), A.I.M.

SYNOPSIS

The characteristic microstructure of peeled samples of whiteheart malleable cast iron is described, and it is shown that the prime cause of peeling is the inward penetration of sulphur from the annealing ore or from sulphur-containing gases in the annealing atmosphere. The depth of the peeled layer depends upon (*a*) the decarburizing power of the annealing medium, (*b*) the actual sulphur content of the ore, (*c*) the manganese/sulphur ratio in the metal, (*d*) the initial carbon content of the metal, (*e*) the silicon content of the metal, (*f*) the annealing time, and (*g*) the annealing temperature. Peeling is demonstrated to be a complex phenomenon involving decarburization, sulphurization, and the inward penetration of oxygen ; desulphurization may also occur. The diffusion processes involved are discussed and related as far as the experimental evidence permits.

THIS statement is made to indicate and emphasize, particularly to those concerned with the commercial production of whiteheart malleable cast iron, the feature which the author considers to be of most practical importance in connection with the formation of peel. A high sulphur content in the annealing ore or in the annealing atmosphere is primarily responsible for the occurrence of the phenomenon, as it is known in commercial production, and if this high sulphur content can be avoided then trouble due to peeling will not occur.

All the micrographs show specimens etched in a solution of 4% picric acid in alcohol. Magnifications of 60 dia. were obtained by means of a 16-mm. achromatic objective, N.A. 0·28, without an eyepiece. Magnifications of 600 dia. were obtained by means of a 4-mm. apochromatic objective, N.A. 0·95.

TYPICAL PEELED STRUCTURES

It is convenient to commence the consideration of peeling in whiteheart malleable iron by discussing the microstructure of a typical commercial peeled sample. Figs. 1 to 4 are micrographs taken from such a sample. Fig. 1 shows the structure at the point where the peel has become detached and Fig. 2 shows the structure at a point where the peel still adhered to and formed part of the casting. It will be noted that the succession of phases and structures is the same in the cases of both the adherent and the detached peel. The actual layer of peel possesses a structure with easily recognizable characteristics which vary in degree from sample to sample. In the outer zones of the peel the structure usually consists of ferrite with coarse globules of an oxide phase which become finer toward the junction of the peel and the sound casting. At the same time, on approaching this interface the oxide phase decreases in amount and tends to be arranged in a network associated with iron sulphide (*see* Fig. 3) which increases in quantity until at the interface an almost continuous network of sulphide exists. Corresponding with the decreasing amount of the oxide phase and the increasing amounts of iron sulphide, there usually exist progressively increasing amounts of coarsely laminated pearlite. Proceeding towards the centre of the casting, and immediately beyond the concentration of iron sulphide, there is usually a sharp change in microstructure to that of normal whiteheart malleable. In the example illustrated in Figs. 1 and 2 the structure becomes completely ferritic beyond the sulphide layer ; Fig. 4 illustrates the structure of the same sample as shown in Fig. 2, in the region of the sulphide concentration, but at a higher magnification, and the oxide phase in a network pattern, the coarsely laminated pearlite, and the sulphide concentration can be clearly seen. The oxide phase in the peeled zone is very dark grey in colour and is partly translucent. It would appear to be a silicate, and is probably fayalite ($2FeO.SiO_2$). It would seem probable that the peeled zones become detached from the main body of the casting by virtue of the local weakening effect of the sulphide concentration.

In considering the formation of the peeled zone it seems logical to assume that it is related to the inward diffusion of sulphur from the edge of the casting. If the sulphur originates from some external source, there are five main possibilities, namely :

(1) The presence of iron pyrites (FeS_2) in the ore.
(2) The presence of iron sulphide (FeS) in the ore.
(3) The presence of sulphates in the ore.
(4) The presence of sulphur dioxide in the annealing atmosphere.
(5) The presence of hydrogen sulphide in the annealing atmosphere.

It is significant that a serious outbreak of trouble due to the peeling of whiteheart malleable-iron munitions products followed upon the introduction to this country of Riff annealing ore in place of the Cumberland ore previously used. Chemical analyses from a number of samples of both ores showed conclusively that the sulphur content of Riff ore was considerably higher than that of Cumberland ore.

No case of peeling has yet been brought to the attention of the author in which the sulphide concentration between the peel and the core of the casting was absent. On account of this it was thought that

Manuscript received 26th April, 1947.
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the sulphide concentration was essential for peeling to occur, and this theory formed the basis upon which the experimental work recorded in this paper was conceived.

EXPERIMENTAL WORK

From the description which has already been given of a typical peeled sample, and from a brief review of the literature, it is obvious that the formation of peel is related to the formation of a sulphide concentration some distance from the surface of the casting. The first object of the present author's experimental work was to demonstrate that this sulphide concentration was due to the use of ores of high sulphur content. Work was carried out to study the progressive formation of the peeled layer during the annealing treatment. This was followed by a study of the various factors which encourage or inhibit peeling. Much of this work was carried out by annealing in conventional ore mixtures, but it was ultimately extended to cover annealing and decarburization under gaseous conditions in the absence of ore. Many subsidiary experiments were carried out and are described in the appropriate places in what follows.

INFLUENCE OF SULPHUR CONTENT OF ORE ON PEEL FORMATION

Four standard $8 \times 1 \times \frac{3}{8}$-in. bend-test bars were each annealed in ores of different sulphur contents. The various sulphur contents were obtained by the addition of calculated quantities of ground iron pyrites (FeS_2) to samples of ore consisting of six parts of used ore to one part of new. The chemical composition of this ore mixture before the addition of pyrites was :

Fe	24·54%
FeO	57·79%
Fe_2O_3	5·79%
SiO_2	7·90%
Al_2O_3	0·05%
CaO	0·56%
MgO	0·48%
Sulphur	0·18%

The chemical composition of the test-bars was :

Total carbon	3·15%
Silicon	0·61%
Manganese	0·15%
Sulphur	0·232%
Phosphorus	0·048%

The iron pyrites was added to three samples of ore to give the range of sulphur contents shown in Table I below :

Table I

SULPHUR CONTENTS OF ORE MIXTURES

Bar No.	Amount of Sulphur Added as FeS_2, %	Total Sulphur Content of Ore, %
1	Nil	0·18
2	0·2	0·38
3	0·4	0·58
4	0·8	0·98

The iron pyrites was ground until it passed through a 60-mesh B.S.I. sieve and was then thoroughy mixed with the ore. The test-bars were packed separately in specially made annealing boxes which were then well luted with raw red sand and finally coated over with a refractory cement. The ore/metal ratio was approximately 1 : 1 by weight.

All the bars were annealed in the same furnace, which was electrically heated and was fitted with an automatic temperature control. No attempt was made to control the furnace atmosphere, which was the natural air atmosphere. The annealing cycle was as follows :

	Time, hr.
Room temperature to 1000° C.	30
Soaking time at 1000° C.	100
1000° C. down to 800° C.	12
800° C. to room temperature	18*
Total annealing time	160

* Approximately

After annealing, the bars were bent to fracture in a standard bend-test machine. Bar 1 showed no evidence of peeling. Bar 2 showed traces of a very thin peel, while bars 3 and 4 showed well-developed peeling. In both these latter cases the thickness of the peel was of the order of 0·06 in. These observations were confirmed by micro-examination.

The structure at the edge of bar 1, which was annealed in ore containing 0·18% of sulphur and to which no addition of iron pyrites was made, is illustrated in Fig. 5. It will be seen that the test-bar had the usual thin oxidized rim, typical of whiteheart malleable cast iron, followed by a ferrite band and then increasing amounts of pearlite and temper carbon. No peeling effect is discernible. The structure at the edge of bar 2 was very similar to that of bar 1, with the exception that the oxidized rim was slightly deeper.

The structure at the edge of bar 3, which was annealed in ore to which an addition of 0·4% of sulphur, as iron pyrites, had been made to a given total sulphur content in the ore of 0·58%, is shown in Fig. 6. This test-bar possessed a structure very similar to that of the commercial sample illustrated in Fig. 2, the same sequence of structures being present in the peel.

Bar 4, which was annealed in ore containing sufficient iron pyrites to raise the total sulphur content to 0·98%, possessed a structure almost identical with that of bar 3, illustrated in Fig. 6. The further addition of 0·4% of sulphur above 0·58% did not appear to have any marked effect.

These experiments proved that peeling could be produced by increasing the sulphur content of the ore, in this case with iron pyrites. It would appear that the critical sulphur content in the ore above which peeling can occur lies between 0·3% and 0·6%, for metal of the composition given above and annealed under similar conditions. This critical sulphur content may be appreciably affected by the annealing time and temperature and the sulphur and manganese contents of the metal. Also, variations in the carbon monoxide, carbon dioxide, and sulphur dioxide contents of the annealing atmosphere and the strength of the ore may have an appreciable effect on the peculiar oxidizing and sulphurizing equilibria required to produce peeling.

Table II
DEPTH OF LAYERS IN ANNEALED TEST-BARS

Bar No.	Depth of Layer Containing Massive Oxide Globules, in.	Total Depth of Peeled Layer, in.
5	0·006 in.	
6	0·020	0·037
7	0·020	0·050
8	0·033	0·075
9	0·008 in.	

Determination of Stage at Which Peel Forms

When the foregoing experiments were completed, it was decided to investigate the stage during the annealing cycle at which the peeling effect was encountered. In order to do this, five standard bend-test bars were annealed in a manner similar to that previously described, according to the following programme :

Bar No.	Sulphur Added as FeS_2, %	Treatment
5	0.5	Annealed for 50 hr. at 1000° C.
6	0.5	Annealed for 75 hr. at 1000° C.
7	0.5	Annealed for 100 hr. at 1000° C.
8	0.5	Annealed for 150 hr. at 1000° C.
9	Nil	Annealed for 150 hr. at 1000° C.

Samples of the same ore as was used for the first experiment were employed, and the bars were similarly packed in separate annealing boxes. All the boxes were placed together in the same furnace and the rate of heating used previously was again adopted. When bars 5, 6, and 7 had been annealed at 1000° C. for the required period, the annealing box containing the appropriate test-bar was removed from the furnace and cooled in air. Bars 8 and 9 were allowed to cool in the furnace. The chemical composition of bars 5 to 9 was the same as that of bars 1 to 4. Bars 5, 6, 7, and 8 were packed in ore containing 0·68% of sulphur, and bar 9 was packed in ore containing 0·18% of sulphur.

Table III
CARBON AND SULPHUR CONTENTS OF SKIN MACHINED TEST-BARS

Annealed for 50 hr. (Bar 10)				Annealed for 75 hr. (Bar 11)			
Cut No.	Distance from Surface, in.	Total Carbon, %	Sulphur, %	Cut No.	Distance from Surface, in.	Total Carbon, %	Sulphur, %
1	0·010	0·23	0·196	1	0·020	0·25	0·175
2	0·025	0·31	0·202	2	0·035	0·37	0·191
3	0·035	0·33	0·193	3	0·050	0·40	0·191
4	0·045	0·41	0·203	4	0·070	0·53	0·209
5	0·055	0·44	0·214	5	0·090	0·72	0·203
6	0·070	0·61	0·211	6	0·110	0·86	0·213
7	0·085	0·82	0·212	7	0·130	1·05	0·205
8	0·100	1·00	0·213	8	0·150	1·28	0·205
9	0·120	1·12	0·214	9	0·170	1·37	0·204
10	0·140	1·40	0·219	10	0·190	1·66	0·207
11	0·160	1·56	0·214	11	0·210	1·83	0·208
12	0·180	1·75	0·210	12	0·230	1·77	0·207
13	0·200	1·94	0·206	13	0·250	1·96	0·210
14	0·220	2·04	0·209	14	0·270	2·28	0·208
15	0·240	2·60	0·210	15	0·290	2·25	0·207

After annealing, the bars were bent to fracture in a bend-test machine, and the presence of peeling was detected in bars 6, 7, and 8. The microstructures of bars 5 and 9 were very similar to the structure shown in Fig. 5, and the microstructures of bars 6, 7, and 8 were very similar to the structure shown in Fig. 6. The depth of the zone containing the massive "oxide" globules and the depth of the peeled or oxidized layer were measured in each case, and the figures obtained are tabulated in Table II.

In the series of bars annealed in ore to which an addition of 0·5% of sulphur as iron pyrites had been made, it is seen that the penetration of oxide and sulphide was very slight in the first 50 hr. (bar 5); in fact the total depth of oxide penetration was of the same order as that in the bar annealed for 150 hr. in ore to which no sulphur had been added. After the first 50 hr. at the annealing temperature, the depth of the peeled layer increased proportionately with time. The findings of this experiment suggest that ore containing much higher sulphur contents might be more tolerated when annealing light sections requiring a relatively short time than when annealing heavy sections requiring a relatively long time.

Table IV
CARBON AND SULPHUR CONTENTS OF SKIN-MACHINED TEST-BARS

Cut No.	Annealed for 100 hr. (Bar 12)			Annealed for 125 hr. (Bar 13)	
	Distance from Surface, in.	Total Carbon, %	Sulphur, %	Total Carbon, %	Sulphur, %
1	0·020	0·16	0·199	0·24	0·132
2	0·040	0·23	0·206	0·27	0·161
3	0·060	0·28	0·215	0·35	0·197
4	0·080	0·37	0·217	0·46	0·236
5	0·100	0·57	0·221	0·60	0·281
6	0·120	0·81	0·212	0·67	0·270
7	0·140	0·96	0·210	0·77	0·266
8	0·160	1·14	0·206	0·96	0·253
9	0·180	1·36	0·207	1·19	0·228
10	0·200	1·31	0·208	1·37	0·265
11	0·220	1·51	0·209	1·44	0·222
12	0·240	1·66	0·209	1·47	0·221
13	0·260	1·70	0·214	1·50	0·209
14	0·280	1·96	0·210	1·63	0·208
15	0·300	1·83	0·212	1·87	0·208

Cut No.	Annealed for 150 hr. (Bar 14)			Annealed for 150 hr. (Bar 15 (blank bar))	
	Distance from Surface, in.	Total Carbon, %	Sulphur, %	Total Carbon, %	Sulphur, %
1	0·020	0·10	0·069	0·05	0·157
2	0·040	0·21	0·111	0·18	0·190
3	0·060	0·36	0·160	0·32	0·181
4	0·080	0·41	0·192	0·36	0·183
5	0·100	0·61	0·247	0·45	0·203
6	0·120	0·73	0·269	0·52	0·199
7	0·140	0·96	0·278	0·74	0·205
8	0·160	1·05	0·263	0·87	0·203
9	0·180	1·25	0·228	0·96	0·204
10	0·200	1·47	0·207	1·09	0·189
11	0·220	1·53	0·207	1·22	0·192
12	0·240	1·53	0·207	1·32	0·200
13	0·260	1·59	0·214	1·44	0·197
14	0·280	1·62	0·197	1·49	0·191
15	0·300	1·75	0·202	1·51	0·191

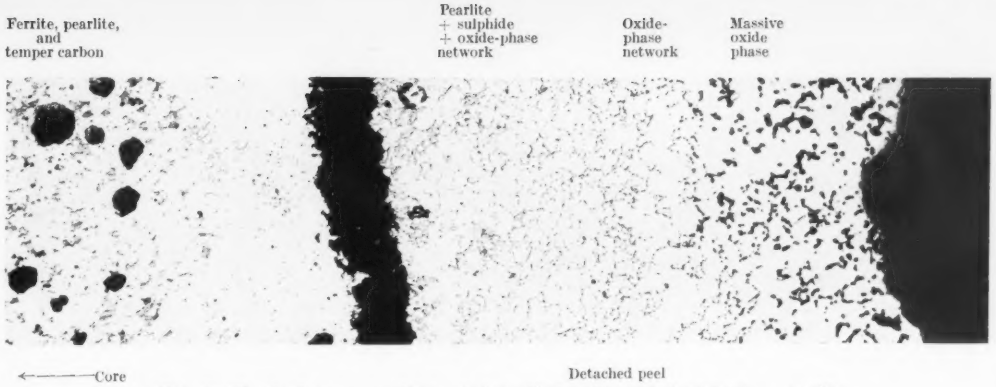

Fig. 1—Typical commercial peeled casting with detached peel × 60

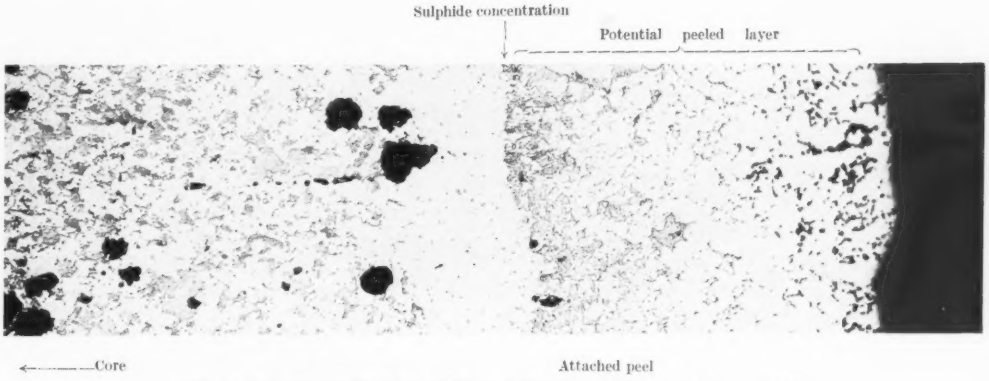

Fig. 2—Same sample as in Fig. 1, showing attached peel × 60

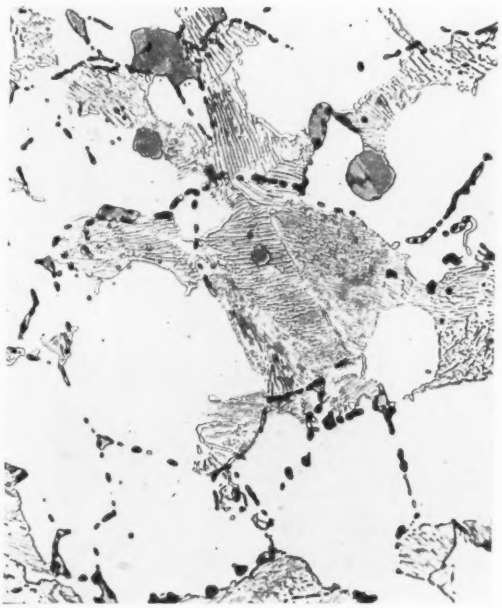

Fig. 3—Appearance of oxide-sulphide phase usually seen at the grain boundaries in the peeled layer × 600

Fig. 4—Same sample as in Fig. 2, showing sulphide concentration at the junction of the peel and the sound core × 600

[*Bernstein*

[*To face p.* 22

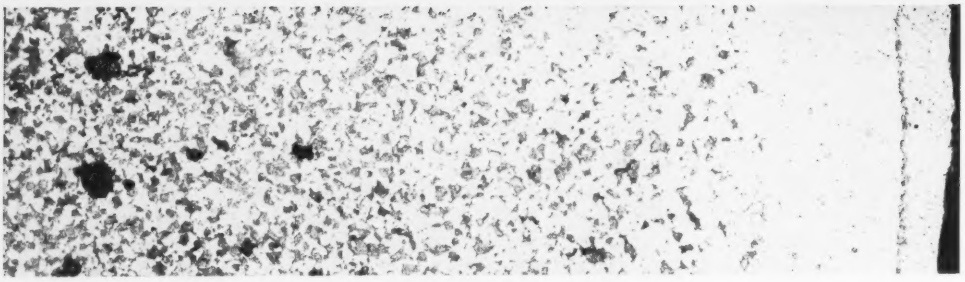

Fig. 5—Structure at edge of bar 1, annealed in ore containing 0·58% of sulphur × 60

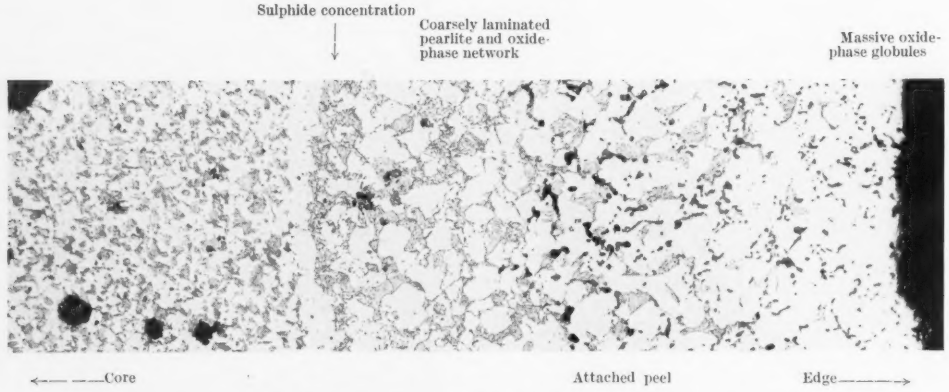

Fig. 6—Structure at edge of bar 3, annealed in ore containing 0·58% of sulphur × 60

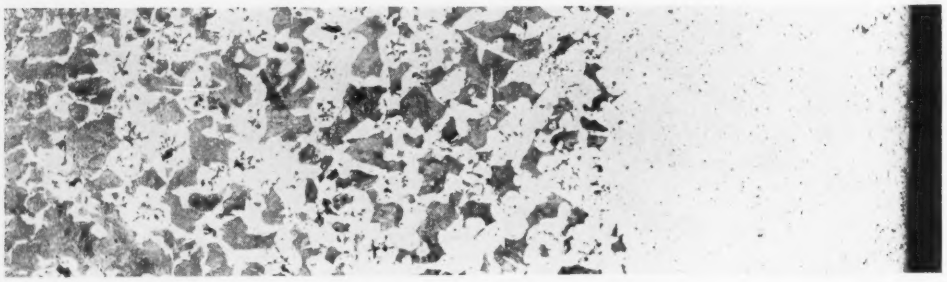

Fig. 7—Structure at edge of bar 27, annealed for 125 hr. at 1000° C. in ore containing 1% of sulphur × 60

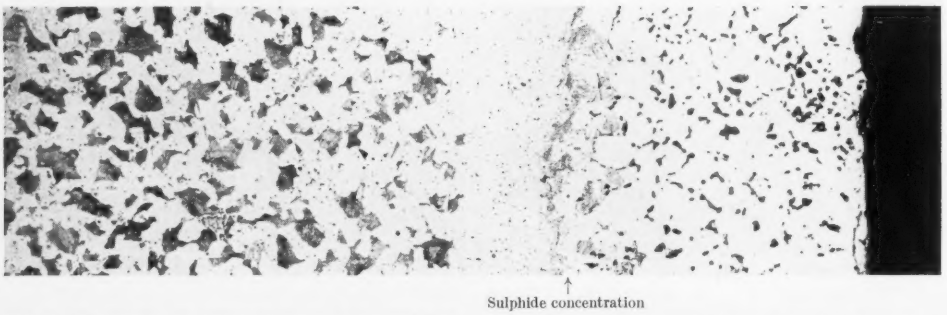

Fig. 8—Structure at edge of sample containing 0·69% of silicon, annealed in ore containing 1% of sulphur × 60

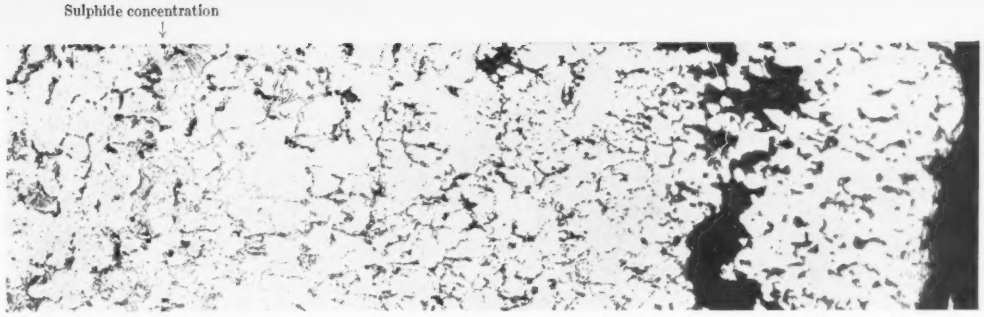

Fig. 9—Structure at edge of sample containing 1·72% of silicon, annealed in ore containing 1% of sulphur × 60

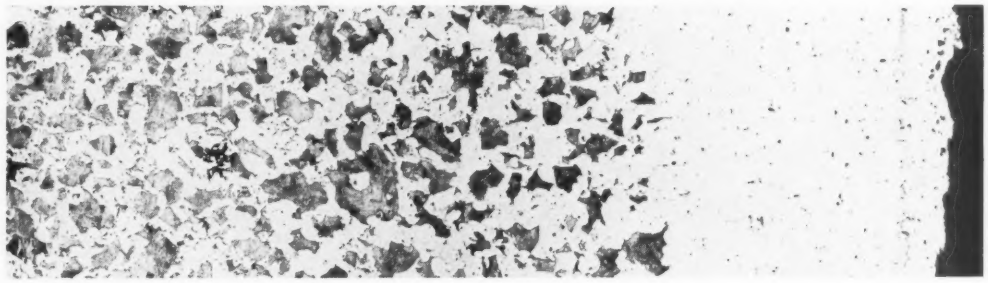

Fig. 10—Structure at edge of sample with the same silicon content as that shown in Fig. 8 and annealed in ore containing 0·4% of sulphur × 60

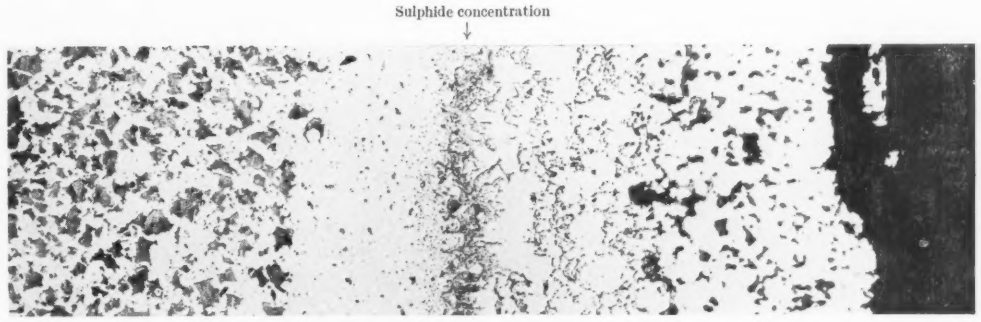

Fig. 11—Structure at edge of sample containing 2·35% of total carbon, 1·29% of silicon, and 0·14% of manganese, annealed in ore containing 1% of sulphur × 60

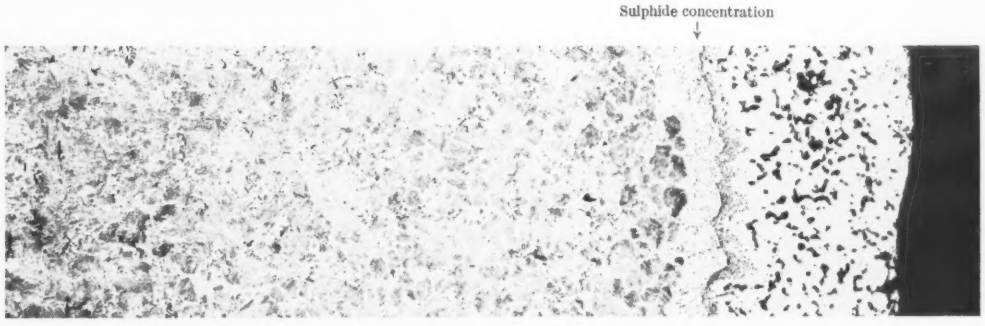

Fig. 12—Structure at edge of sample with same base composition as that shown in Fig. 11, but with a higher manganese content. Annealed in ore containing 1% of sulphur × 60

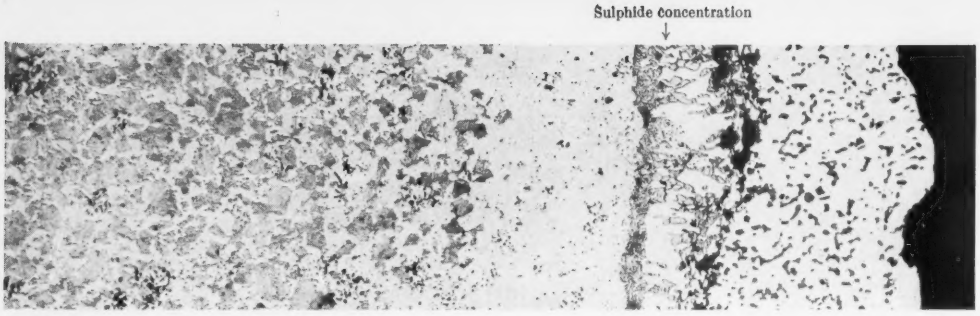

Fig. 13—Structure at edge of sample with same composition as that shown in Fig. 11, annealed in ore containing 0·4% of sulphur × 60

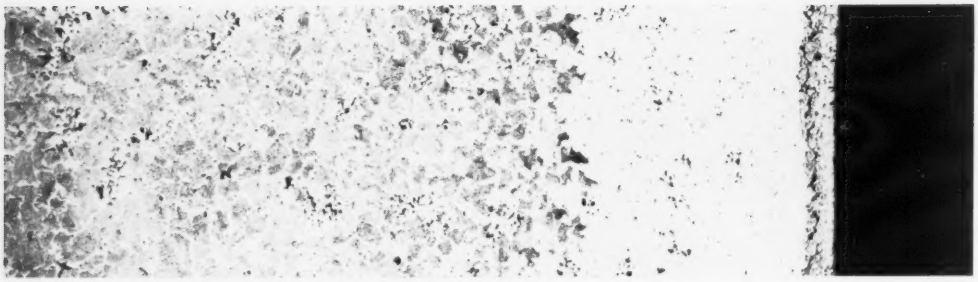

Fig. 14—Structure at edge of sample with same base composition as that shown in Fig. 13, but with a manganese content of 0·39%. Annealed in ore containing 0·4% of sulphur × 60

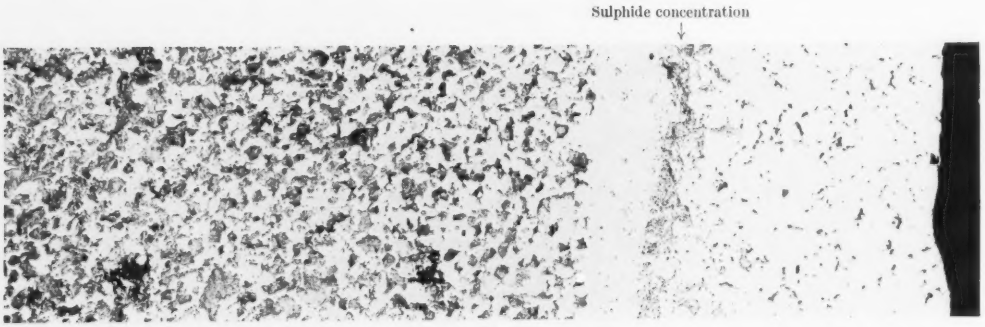

Fig. 15—Structure at edge of sample containing 3·25% of total carbon, 0·57% of silicon, 0·29% of manganese, and 0·21% of sulphur, annealed in ore containing 1% of sulphur × 60

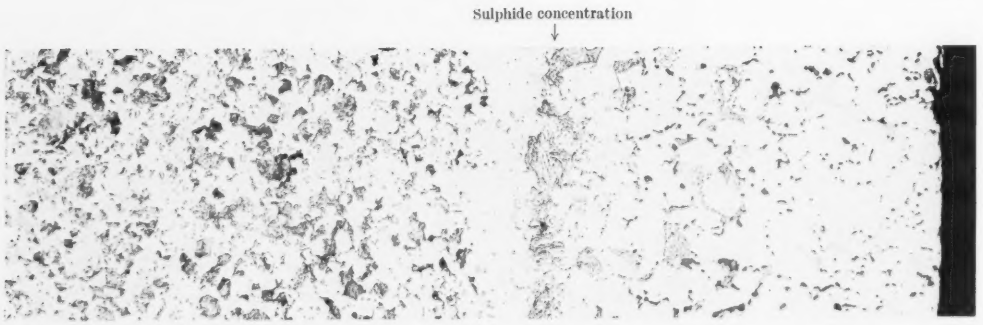

Fig. 16—Structure at edge of sample with same base composition as that shown in Fig. 15, but with 0·334% of sulphur. Annealed in ore containing 1% of sulphur × 60

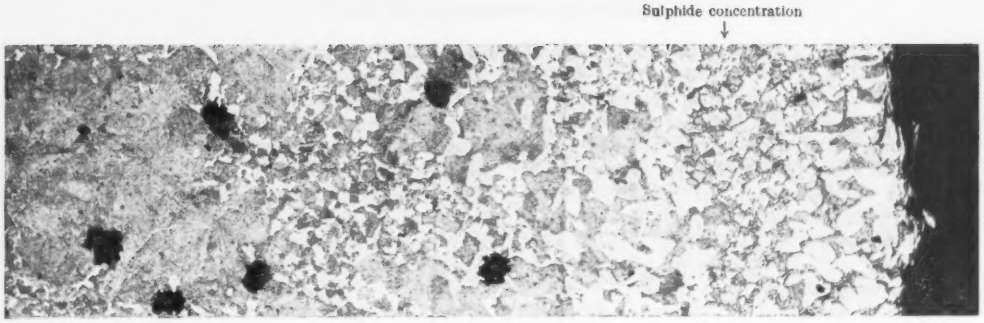

Fig. 17—Depth of peel at edge of bar 28, annealed at 1000° C. in an atmosphere containing 5% of CO_2 × 60

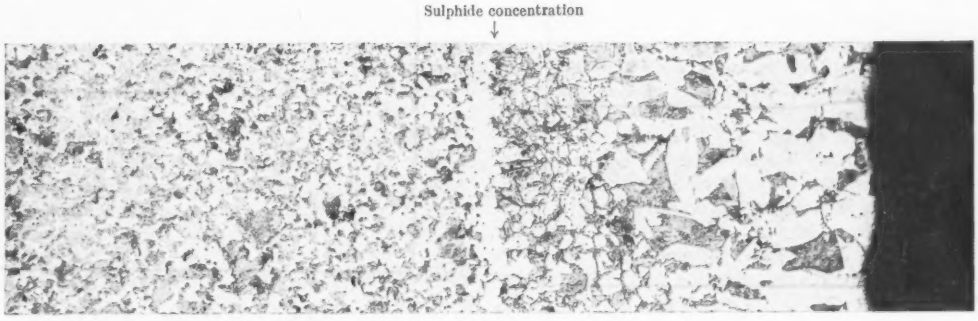

Fig. 18—Depth of peel at edge of bar 29, annealed at 1000° C. in an atmosphere containing 10% of CO_2 × 60

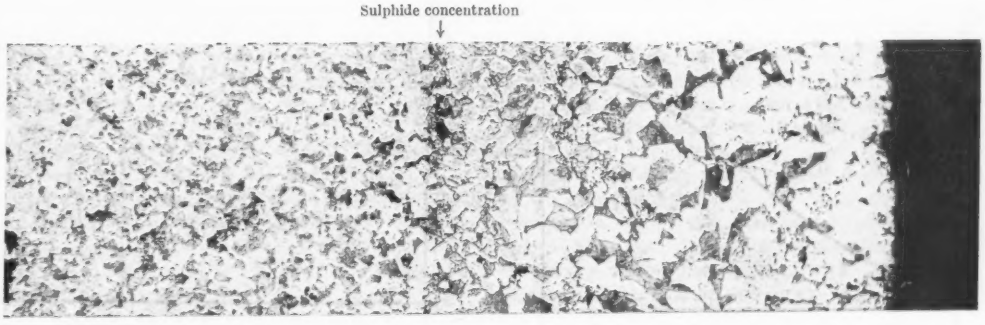

Fig. 19—Depth of peel in bar 31, annealed in a gaseous mixture containing a small amount of water vapour × 60

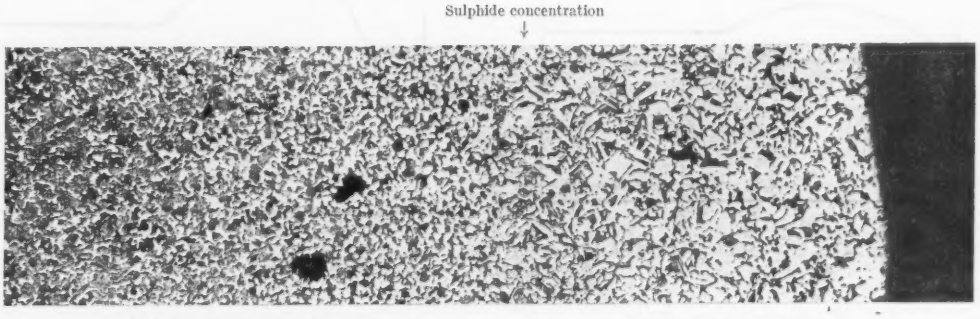

Fig. 20—Structure at edge of bar air-cooled from 1000° C. × 60

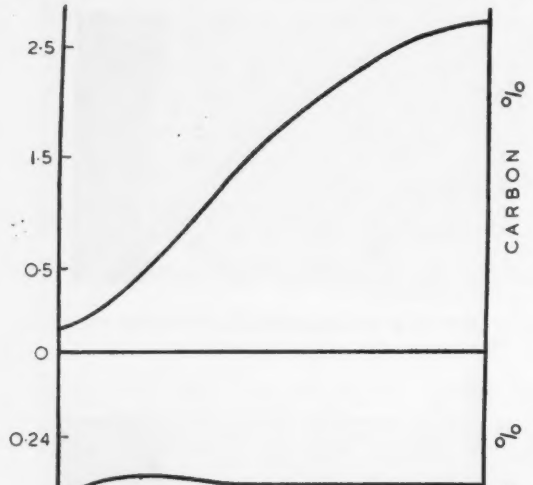

Fig. 21—Annealed for 50 hr.

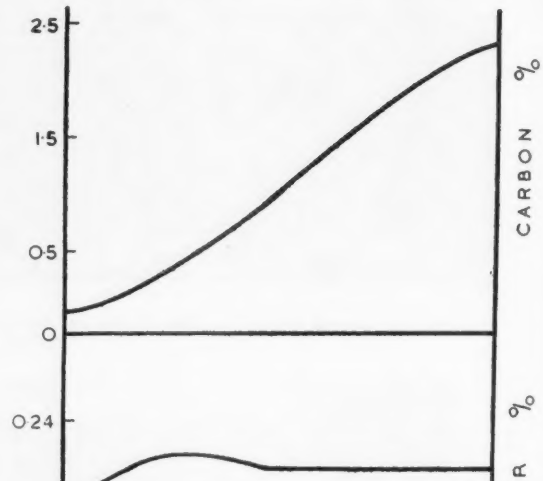

Fig. 22—Annealed for 75 hr.

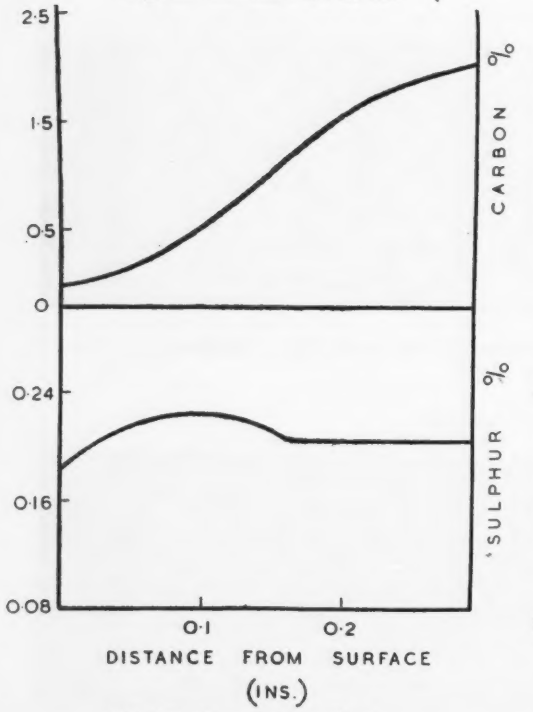

Fig. 23—Annealed for 100 hr.

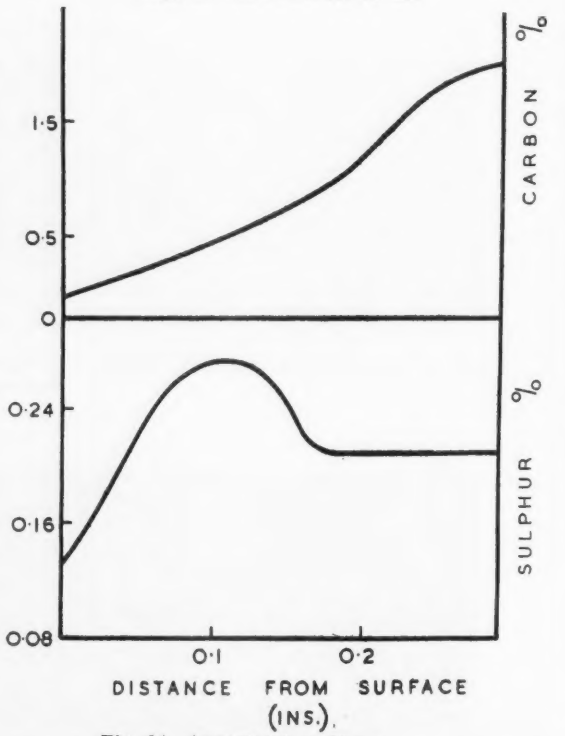

Fig. 24—Annealed for 125 hr.

Figs. 21 to 24—Diffusion of carbon and sulphur in test-bars annealed at 1000° C. in ore containing 0·68% of sulphur

It is obvious from the experiments described so far that the presence of sulphides in the annealing ore drastically increases the tendency for the penetration of oxide into the metal. The metallographic evidence shows iron sulphide in the peel to be closely associated with iron oxide (possibly FeO). Oberhoffer[1] reports that iron sulphide (FeS) and iron oxide (FeO) form a eutectic melting at 940° C., and that the heating of samples containing FeS in an oxidizing atmosphere causes the formation of this FeS–FeO eutectic. Thus it may be seen that the iron-sulphide/iron-oxide complexes seen in peeled samples may be molten at the annealing temperature. The high mobility of this liquid would help to explain the remarkably deep penetration of oxide and sulphide in peel. Wohrman[2] heat-treated a pure iron–sulphur alloy containing 1·05% of sulphur at 960° C. and 990° C., and found abnormal penetration of oxide even after only 1–2 hr. He suggests that a more or less continuous sulphide network must have served to carry the oxide into the samples. The oxide and sulphide were present as the FeS–FeO eutectic. Preece[3] says that during the heat-treatment of steel in furnace atmospheres containing sulphurous gases at temperatures above 900° C., the sulphide forms a molten oxide–sulphide complex penetrating along the grain boundaries.

Considering these known facts, it seems possible that an iron sulphide in the ore might be partially oxidized to give sulphur dioxide, which in turn causes sulphurization of the metal, giving a concentration of iron sulphide. This iron sulphide causes a rapid absorption of oxygen as FeO, the two combining to form a molten sulphide/oxide-phase complex which readily penetrates into the metal. The structure illustrated in Fig. 3 shows the type of oxide-phase/sulphide complex usually seen at grain boundaries in the peeled layer.

DIFFUSION OF CARBON AND SULPHUR DURING THE FORMATION OF PEEL

For investigation of the diffusion of carbon and sulphur during the formation of peel, six 0·875-in. dia. bars 10 in. long (bars 10 to 15) were used in place of the standard bend-test bars of the previous experiments. The as-cast skin of the bars was ground off until the surface of each bar was perfectly clean. The bars were then packed in ore of strength six parts of used ore to one part of new ore, in separate annealing boxes, the ore/metal ratio being approximately 1 : 1 by weight. The sulphur content of the ore was raised to 0·68% in each case, by the addition of 0·5% of sulphur as iron pyrites, with the exception of the ore for bar 15, to which no sulphur was added. The test-bar annealed in this ore, which contained 0·18% of sulphur, was used as a " blank."

The chemical composition of the bars was as follows :

Total carbon	3·35%
Silicon	0·54%
Manganese	0·19%
Sulphur	0·195%
Phosphorus	0·08%

When packed in ore in their separate annealing boxes, the bars were all placed in the same furnace, which was heated to 1000° C. in 30 hr. After 50 hr. at 1000° C., one of the bars (bar 10) was removed in its annealing box and allowed to cool in air. After a further 25 hr., the next bar (bar 11) was removed and allowed to cool in air. Two further bars (bars 12 and 13) were removed after being soaked at 1000° C. for 100 hr. and 125 hr., respectively. These were also allowed to cool in air. The remaining two bars (bars 14 and 15) were allowed to cool in the furnace after a total of 150 hr. at 1000° C.

The bars were then skin-machined, leaving only sufficient unmachined for subsequent micro-examination. The turnings were collected and the depth of the cut noted. Skin-machining was continued, 14 further cuts being removed and the turnings obtained from each cut collected separately. The total carbon and sulphur contents of each of the samples of turnings were then determined, and the figures obtained for each bar were plotted graphically against the distance from the surface. The results obtained are shown in Figs. 21 to 26, and the actual figures are given in Tables III and IV. It will be seen that the bars annealed for 125 hr. and 150 hr. in the ore of high sulphur content developed a marked segregation of sulphur in the region 0·1 in.–0·2 in. from the surface. Microspecimens were cut from the unmachined portions of the bars to represent transverse sections, and examination showed that the distribution of the iron sulphide corresponded, as one would anticipate, with the analytical figures.

The microstructures are not reproduced in this paper, but the structures of bars 10, 11, and 15 were very similar to the structure shown in Fig. 5, except that less decarburization had occurred in the former bars. The structures at the edges of bars 13 and 14 were very similar to that shown in Fig. 6, and the structure of the edge of bar 12 was approximately midway between those illustrated in Figs. 5 and 6 with respect to the depth of penetration of sulphide and the oxide phase.

It is interesting to note that machining did not cause the potentially peeled layer to strip off, as sometimes does occur. However, precautions were taken to avoid the separation of peel during machining by taking off very shallow cuts with a very sharp tool. An experiment carried out by the author showed that the edge effect described as the potential peeled edge or layer and illustrated in Figs. 2 and 6 definitely indicates that peeling would occur and that it does so

Table V
SULPHUR CONTENTS OF ORES USED IN ANNEALING

Annealing Time	Sulphur, %		Sulphur Loss on Annealing, %
	Before Annealing (calculated)	After Annealing	
50 hr. (bar 10)	0·68	0·426	0·254
75 hr. (bar 11) ...	0·68	0·390	0·290
100 hr. (bar 12) ...	0·68	0·460	0·220
125 hr. (bar 13) ...	0·68	0·473	0·207
150 hr. (bar 14) ...	0·68	0·485	0·195
150 hr. (bar 15) ...	0·18	0·159	0·021

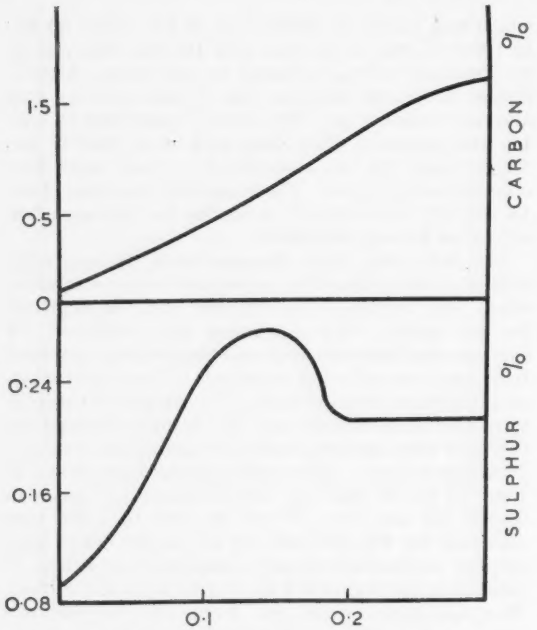

Fig. 25—Diffusion of carbon and sulphur in test-bars annealed for 150 hr. at 1000° C. in ore containing 0·68% of sulphur

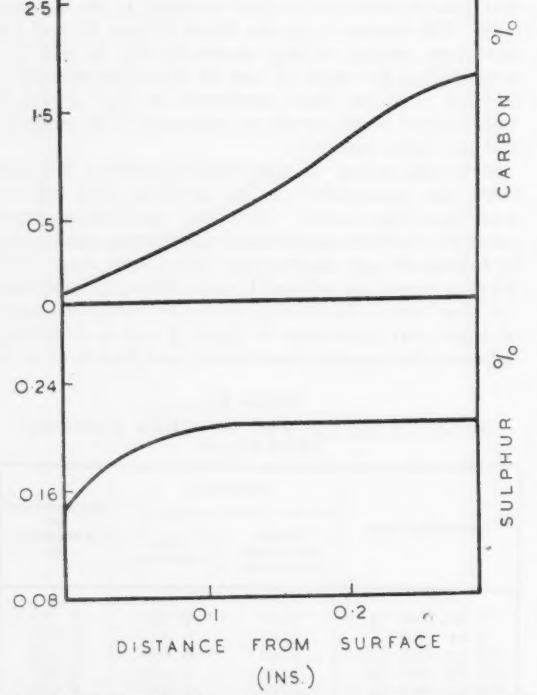

Fig. 26—Diffusion of carbon and sulphur in test-bars annealed for 150 hr. at 1000° C. in ore containing 0·18% of sulphur

at the point of highest concentration of iron sulphide, if mechanical stress is applied.

An examination of the results given in Tables III and IV and Figs. 21 to 26 reveals some interesting features. The sulphur content of the bars annealed in sulphurous ores shows a progressive drop near the surface with increasing annealing time. Corresponding with this is the gradual increase in sulphur content at a point some distance from the surface. This sulphur concentration corresponds with that seen at the junction of the peel and sound metal. It is obvious from the curves for the 125-hr. and the 150-hr. anneal that the sulphur distribution is not the result of continuous diffusion of sulphur from the ore into the metal, as the extreme outer edge of the samples contains much less sulphur than the original white iron. If it were a case of continuous diffusion of sulphur from the edge to the centre, one would expect a maximum sulphur content at the surface, decreasing towards the centre.

When these results are plotted in graphical form and analysed planimetrically with due consideration to the relative masses of metal involved, it can be shown that the total amount of sulphur present in the area of concentration in excess of that originally present is greater than the amount which has been eliminated from the outer layers of the peel. This indicates that the metal has absorbed a certain amount of sulphur from the ore. The sulphur contents of the used ores (*see* Table V) show that sulphur has been lost by the ore during these experiments.

The total carbon determinations, the results of which are also shown in Tables III and IV and Figs. 21 to 26, do not reveal any particularly interesting features beyond a slight concentration of carbon at the limit of the sulphur segregation.

EFFECT OF ANNEALING IN ORES OF VARIOUS STRENGTHS BUT CONSTANT SULPHUR CONTENT UPON PEEL FORMATION

The strength of the annealing ore is one of the factors which has been said to influence the formation of peel. There appears to be some disagreement as to the advisability of using strong ore, although the idea was expressed in the earlier literature that, as peel is in some way related to oxidation, the use of strong ore would give rise to peel. However, at the time of the outcrop of trouble due to peeling experienced during the recent war, an opinion was expressed by some that the use of strong ore could be of assistance in avoiding peel. It was to determine whether, and if so to what extent, the formation of peel is influenced by the strength of the ore that the next experiments to be described were carried out.

Two experiments were performed in exactly the same manner, the first with a total sulphur content in the ore of 0·7% and the second with a total sulphur content in the ore of 1%. The results of these two experiments agree very closely, and hence a description of the procedure, details, and results of only the second experiment will be necessary here.

Eight different ore mixtures and eight standard bend-test bars from the same cast were used. The analysis of the test-bars was as follows :

Total carbon	2·85%
Silicon	0·50%
Manganese	0·10%
Sulphur	0·191%
Phosphorus	0·064%

The bars were packed as previously described, in annealing ore to which appropriate additions of iron pyrites had been made to bring the total sulphur content to 1% in each case. The chemical compositions of the new and the used ores were :

	New Ore	Used Ore
Fe, %	Nil	41·78
FeO, %	2·37	42·53
Fe_2O_3, %	90·00	3·76
CaO, %	0·89	0·88
SiO_2, %	3·62	8·79
Sulphur, % ...	0·146	0·494*

* Before addition of FeS_2

It will be seen that the total sulphur content of the used ore was considerably higher than that of the new ore. Thus the amount of pyrites added to the ore increased with increasing ore strength. In this manner the sulphur content of the ore was maintained at a constant value for the different ore strengths. The actual ratios of new to used ore which were used are given in Table VI.

The labelled annealing boxes were placed together in an electrically heated furnace, the temperature of which was raised from room temperature to 1000° C. in 30 hr. The temperature was maintained at 1000° C. for 120 hr. and was then allowed to fall to room temperature in 36 hr. The test-bars were then removed from the annealing boxes and bent to fracture in a bend-test machine. Examination of the fractures showed that all the bars exhibited peeling, the extent of which differed considerably from bar to bar. Microspecimens were cut from the bars, close to the fractures, and the distance between the surface and the iron sulphide concentration was measured on each microspecimen. The results obtained are included in Table VI.

The most important features indicated by these results is that peeling occurred in every case, even with the richest ore. It will be seen that, starting with 100% new ore, the depth of the peeled layer increases with a decrease in the richness of the ore until a maximum is reached over the range of ore strengths 3 : 1, 4 : 1, and 6 : 1, used to new ore. Further reduction in the strength of the ore causes a reduction in the thickness of the peeled layer. Similar results were obtained with ore containing 0·7% of sulphur.

Table VI
ORE STRENGTHS USED AND DEPTH OF SULPHIDE

Bar No.	Ore-Strength Ratio		Depth of FeS Concentration, in.	
	Used	New	Maximum	Minimum
16	0	1	0·040	0·018
17	1	1	0·044	0·018
18	2	1	0·049	0·033
19	3	1	0·065	0·012
20	4	1	0·058	0·032
21	6	1	0·063	0·046
22	8	1	0·027	0·026
23	10	1	0·036	0·023

Table VII
ADDITIONS OF SULPHUR TO ORES USED IN ANNEALING

Bar No.	Addition of Sulphur as FeS_2, %	Total Sulphur Content of Ore, %
24	Nil	0·18
25	0·2	0·38
26	0·4	0·58
27	0·8	0·98

It is felt that this experiment confirms the point of view that sulphur in the annealing ore or annealing atmosphere is the prime cause of peeling. If the sulphur content of the annealing ore is high, then variations in ore strength are of little consequence. When, however, the sulphur content of the new ore is low, then an increase in the ratio of new ore to used ore may prevent the occurrence of peel by reducing the average sulphur content of the aggregate. This would explain why, in certain cases, peeling has been avoided by increasing the amount of new ore used. It should be pointed out that this method of avoiding peeling will be efficacious only when it results in a reduction of the sulphur content of the ore, and if this condition is not fulfilled an increase in ore strength may aggravate the peeling. These findings have subsequently been confirmed on many occasions, and by gaseous-decarburization experiments to be described later.

THE PREVENTION OF PEELING BY MANGANESE

A consideration can now be made of the factors which may inhibit peeling. It would appear that the formation of iron sulphide at the surface of the casting is a link in the sequence of reactions producing peeling. Therefore, if this step can be prevented when annealing in a sulphur-containing ore, the whole process of peeling may be arrested.

If an excess of manganese is present in the casting, owing to the marked affinity of manganese for sulphur it would be expected that manganese sulphide would be formed instead of iron sulphide. Manganese sulphide does not appear to diffuse at all readily in iron. It has a much higher melting point than iron sulphide, and oxide/manganese-sulphide complexes have much higher melting points than the similar iron sulphide complexes.[4] These considerations suggest that if a manganese content in the metal in excess of that required to balance the sulphur was used, then peeling might be inhibited even when annealing in ore of high sulphur content.

It has recently been shown by Palmer[5] that the amount of manganese required to balance the sulphur content in malleable iron is given by the formula : Manganese (%) = 1·7 × sulphur (%). Any excess of manganese over this figure will be dissolved in the metal and will be available for combination with sulphur coming from an external source, for example from the ore during annealing.

In order to check this point, four test-bars were annealed under conditions identical with those

described in the first experiments. The annealing ore was made up as shown in Table VII, using a ratio of six parts of used ore to one part of new ore, and the composition of the test-bars was :

Total carbon...	3·14%
Silicon	0·48%
Manganese	0·34%
Sulphur	0·099%
Phosphorus	0·05%

The annealing cycle was as follows :

	Time, hr.
Room temperature to 1000° C.	30
Soaking time at 1000° C.	125
1000° C. down to 800° C.	12
800° C. to room temperature	18
Total annealing time	185

After annealing, the bars were bent in a standard bend-test machine, where they all gave bend values of 180° without any sign of fracture. No peeling was observed in any of the bars, and the absence of peeling was confirmed by micro-examination. The microstructures of all four bars were very similar, and Fig. 7, which illustrates the structure at the edge of bar 27, is typical of the four structures. No evidence of undue oxidation or of sulphur penetration is apparent.

It is felt that this experiment definitely shows that peeling can be avoided with ores containing up to about 1% of sulphur if the metal contains more than 0·35% of manganese and up to about 0·1% of sulphur. It is probable that even higher sulphur contents could be tolerated in the ore with higher manganese contents in the metal. It is also probable that the critical sulphur content permissible in the ore varies with the sulphur content of the metal for a fixed manganese content. Normally the higher the manganese the lower is the sulphur content of the metal, and so the effects of these two elements must be considered together. Since the completion of this experiment, additional confirmation of the beneficial effect of manganese in inhibiting the formation of peel has been obtained from a number of manufacturers of whiteheart malleable iron. Those manufacturers using manganese contents in excess of 0·3% and with relatively low sulphur contents have not experienced peeling, in spite of the use of high-sulphur annealing ores.

The statement that higher manganese contents assist in the elimination of peeling needs to be accompanied by the following reservations. With manganese contents of the order of 0·3–0·4%, the sulphur contents will tend to be lower than those usual in whiteheart malleable iron. This results in a more readily graphitized iron which may give a black fracture in heavy sections. Many users are prejudiced against black-fracture whiteheart malleable iron, although the mechanical properties of this material are frequently superior to those obtained with the more usual material of relatively low manganese and relatively high sulphur contents.

EFFECT OF CHEMICAL COMPOSITION OF THE METAL

In an attempt to obtain more information upon the fundamental principles governing the formation of peel, a study was made of the effect of chemical composition of the metal upon the tendency to peel under standard conditions of annealing. Metal compositions covering whiteheart and blackheart malleable irons, with manganese contents of from 0·14% up to 1·5%, sulphur contents from 0·09% to 0·45%, silicon contents from 0·35% to 1·75%, and covering a wide range of combinations of these elements between these limits for high- and low-carbon irons were studied. The effect of phosphorus was not studied. This work is still not quite complete and it is therefore not intended at this stage to quote the results in their entirety, but some very important points have already emerged. Only a general survey dealing with the salient features of the results will therefore be given.

Throughout this experiment, standard conditions of annealing and packing of test-bars were adhered to in order to facilitate a more accurate comparison. All the annealing was done in the same furnace, which was electrically heated and the temperature of which was accurately controlled with the aid of a programme control unit. The test-bars were cast in the same manner and were packed in ore from one batch. The strength of the ore was 6 : 1 (used to new ore) and the ore/metal ratio was 2 : 1, by weight. The same weight of iron pyrites was added to each of the annealing boxes in which it was required to raise the sulphur content of the ore to 1%. For purposes of comparison, one bar of each composition was packed in ore of lower sulphur content (0·4–0·5%) and annealed at the same time as the corresponding bar in ore of higher sulphur content. The annealing cycle was as follows :

	Time, hr.
Room temperature to 1000° C.	25
Soaking time at 1000° C.	110
1000° C. down to 700° C.	40
700° C. to room temperature	12
Total annealing time	187

After cooling, all the bars were bent to fracture in a bend-test machine, and the fractures were examined for signs of peeling. Microspecimens were cut from the fractured ends of the bars and, after preparation and examination, the depth of the peel, as shown by the depth of penetration of sulphide and oxide, was measured.

SUMMARY OF RESULTS

In what follows, specific examples are given of some of the results obtained arising from variations in the metal composition and in the sulphur content of the ore. These examples are followed by a generalization of the influence of these variables which is considered to be permissible in the light of the experimental evidence.

(1) In an iron of base composition : Total carbon 2·2%, manganese 0·39%, sulphur 0·093%, and in which the silicon contents were increased by stages of 0·3% from 0·5 to 1·7% of silicon, peeling was produced in all bars when the amount of sulphur in the annealing ore was 1%. The amount of peel present increased as the silicon content of the

metal increased. When bars of the same chemical composition were annealed in ore containing 0·4% of sulphur, peeling occurred only when the silicon content exceeded 1·4%. Figure 8 shows the edge of a sample the analysis of which was: Total carbon 2·22%, silicon 0·69%, manganese 0·39%, and sulphur 0·099%, and which was annealed in ore containing 1% of sulphur. Figure 9 shows the edge structure of a similar sample, but with 1·72% of silicon, also annealed in ore containing 1% of sulphur. These two micrographs clearly reveal the influence of silicon in increasing the depth of sulphur penetration. Figure 10 shows the structure of a further sample with the same silicon content as the sample illustrated in Fig. 8 but annealed in ore containing 0·4% of sulphur, and no sulphur penetration is visible. It is important to observe here that high silicon contents will, even in the presence of excess manganese, induce peeling, provided that the sulphur content of the ore is sufficiently high.

(2) Peeling was also formed when test-bars of base composition: Total carbon 2·3%, silicon 1·3%, sulphur 0·093%, and in which the manganese content was increased from 0·14 to 1·47%, were annealed in ore containing 1% of sulphur. The amount of peel decreased as the manganese content of the metal increased. When bars of the same chemical composition were annealed in ore containing approximately 0·4% of sulphur, peeling occurred only when the manganese content of the metal was 0·14% and 0·39%. Manganese contents in the metal above this latter figure completely inhibited the formation of peel. Figure 11 shows the edge structure, with the sulphide concentration, of a sample containing 2·35% of total carbon, 1·29% of silicon, 0·14% of manganese, and 0·093% of sulphur, annealed in ore containing 1% of sulphur. The manganese content here is just sufficient to balance the sulphur content. Figure 12 shows the edge structure of a sample with the same base analysis as that of the sample illustrated in Fig. 11, but with the manganese content raised to 1·4%. This sample was also annealed in ore containing 1% of sulphur, and showed a shallow, but nevertheless marked, sulphur penetration. On comparing Figs. 11 and 12 it will be seen that increasing the manganese reduced the depth of sulphur penetration, but did not prevent it. Figure 13 shows the edge structure of a sample of the same composition as the sample illustrated in Fig. 11 but annealed in ore containing only 0·4% of sulphur. In this case sulphur penetration has occurred, but not so deeply as when ore of 1% was used. Figure 14 shows the edge structure of a further sample of the same base analysis, but with a manganese content of 0·39%, also annealed in ore of 0·4% of sulphur. In this case only a relatively slight excess of manganese has been adequate to inhibit sulphide penetration and peeling.

(3) In an iron containing 3·1% of total carbon, 0·55% of silicon, 0·096% of sulphur, and manganese contents increasing from 0·16 to 1·34%, no peeling occurred when test-bars were annealed in ore containing 0·30% and 1% of sulphur.

(4) In an iron of base composition: Total carbon 3·2%, silicon 0·57%, manganese 0·29%, and in which the sulphur contents were increased from 0·21 to 0·42%, peeling occurred when test-bars were annealed in ore containing 1% of sulphur, the depth of peel increasing with increasing sulphur in the metal. When the sulphur content of the annealing ore was between 0·4 and 0·5%, traces of peeling were observed only when the sulphur content of the metal exceeded 0·3%. Figure 15 shows the edge structure, with sulphide penetration, of a sample containing 3·25% of total carbon, 0·57 of silicon, 0·29% of manganese, and 0·207% of sulphur, and annealed in ore having a sulphur content of 1%. In this case there is insufficient manganese to balance the sulphur in the metal, and even though the silicon content is relatively low, marked sulphide penetration has occurred. A further sample of the same base analysis, but with the sulphur content of the metal raised to 0·334%, is shown in Fig. 16. Comparing Figs. 15 and 16, it is seen that the increase in sulphur content of the metal (with the manganese content remaining constant) has caused an increase in the depth of sulphur penetration.

In these experiments it was again confirmed that peeling occurred only when the annealing ore contained appreciable amounts of sulphur. Under these conditions the following factors were found to influence the severity of the peeling and the depth of sulphide penetration when a wide range of chemical compositions was considered:

(*a*) The actual sulphur content of the ore. (The higher the sulphur content, other conditions being the same, the deeper the peel.)

(*b*) The ratio of manganese to sulphur in the metal. (The higher this ratio, the lower is the tendency to peel.)

(*c*) The silicon content of the metal. (The higher the silicon content the greater is the tendency to peel.)

(*d*) The initial carbon content of the metal. (The higher the carbon content, the lower is the tendency to peel.)

Factors favouring the formation of peel also increase the depth of the peel and sulphide penetration when peeling occurs. Other variables, such as ore strength, time of annealing, and annealing temperature, are known to influence peeling, and these are dealt with elsewhere in this paper.

It should be borne in mind that these conclusions are based on the results of experiments covering much wider variations in chemical composition than are normally experienced in the commercial manufacture of whiteheart malleable iron. For instance, the widest range normally covered is probably 2·8–3·7% of total carbon, 0·3–0·8% of silicon, 0·1–0·5% of manganese, and 0·07–0·4% of sulphur. Day-to-day variations in individual works are, of course, usually confined within still narrower limits. With this in mind, the author is inclined to attach more importance to variations in the manganese and sulphur contents of the metal and the sulphur content of the ore than to variations in the total carbon and the silicon

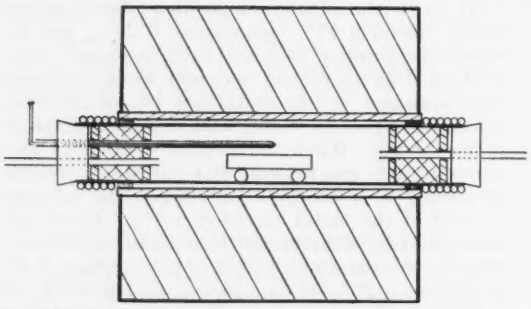

Fig. 27—Diagram of the furnace, showing positions of specimen, thermocouple, and various tubes

contents of the metal. In any given case, all the factors influencing the formation of peel function simultaneously. The general statements made above are qualitative only, and it would, for instance, be impossible at the present stage to indicate a precise safe maximum value for the sulphur content of an ore for annealing metal of a given composition. However, it is felt that for normal practice the sulphur content of the ore should not exceed 0·25%, but if this figure cannot be maintained, freedom from peeling may be achieved by the use of manganese contents in excess of those required to balance the sulphur, according to the formula: Manganese (%) = 1·7 × sulphur (%). This applies only for normal low-silicon contents.

PEELING PRODUCED DURING GASEOUS DECARBURIZATION

As has already been mentioned, the primary cause of peeling is the presence of sulphur in the annealing ore, but the manner in which sulphur is transferred from the ore to the metal simultaneously with the removal of carbon from the metal requires special study. Consideration of the known facts and experimental results relating to the annealing of white iron in ore suggested to the author that the study of peeling could, with valid analogy, be usefully extended to decarburization by the gaseous process.

Decarburization by a number of gas mixtures was therefore studied, but those having most bearing upon the subject under discussion were mixtures of carbon monoxide, carbon dioxide, nitrogen, sulphur dioxide, and water vapour. It can be stated at this stage that peeling can be produced at will by introducing sulphur dioxide into a decarburizing gas mixture containing carbon dioxide, and that the presence of sulphur dioxide is an essential feature for the production of peeling under these conditions. The results obtained from experiments on gaseous decarburization confirm those reported earlier in this paper.

TEST-BARS, APPARATUS, AND PREPARATION OF GAS MIXTURES

The test-bars consisted of machined cylindrical white-iron bars 0·825 or 0·600 in. in dia. and the length of which depended upon the particular experiments.

The annealing furnaces were of the horizontal cylindrical-tube type and consisted of sillimanite tubes, 30 in. long and $2\frac{3}{4}$ in. in dia., on which a Brightray wire heating element was wound over a length of 20 in. A silica tube, 36 in. long and $2\frac{1}{4}$ in. in dia., fitted with rubber bungs carrying small-diameter silica tubing to serve as a thermocouple sheath and to facilitate the entry and exit of the gas mixtures, was placed inside the sillimanite tube. The rubber bungs, which were sealed-in with wax, were protected by silica baffles separated by asbestos wool and were water-cooled. The positions of the specimen and the various tubes are diagrammatically shown in Fig. 27. The temperatures of the furnaces were automatically controlled to within ± 2°C.

For these experiments commercial carbon dioxide, carbon monoxide, and nitrogen were used. Precautions were taken to remove oxygen, carbon monoxide, and carbon dioxide from the particular gases by means of chromous chloride solution, ammoniacal cuprous chloride solution, and either 45% potassium hydroxide solution or soda asbestos, respectively. Water vapour was removed by anhydrous calcium chloride, followed by phosphorus pentoxide contained in a number of tall drying towers. Sulphuric acid was used to remove any ammonia vapour evolved from the cuprous chloride solution. The purified gases were mixed in the required proportions and were divided into two portions, to one of which the required amount of sulphur dioxide was added.

The rate of flow of the gas mixtures was controlled between 100 and 200 c.c./min., dependent upon the size of the test-bar being annealed, and the direction of flow of the mixtures was reversed at regular intervals in order to ensure that uniform decarburization was effected. All air in the apparatus was displaced by the appropriate gas mixtures before commencing each experiment.

The effect of annealing white iron in a considerable number of gas mixtures, with and without sulphur dioxide, was studied over the range of gas compositions given below:

Carbon dioxide ...	... 0– 20%
Carbon monoxide ...	... 30–100%
Nitrogen	... 0– 50%
Water vapour ...	... 0– 2%*
Sulphur dioxide ...	... 0– 2%*

* Approximate values

Tests were carried out for 50–200 hr. at temperatures of 950°, 1000°, and 1050° C. When considered as three main groups the results of all the experiments may be discussed together. The three groups are those in which were used, (i) dry carbon-monoxide/carbon-dioxide mixtures, (ii) dry carbon-monoxide/carbon-dioxide/nitrogen mixtures, in which nitrogen replaced part of the carbon monoxide, and (iii) carbon-monoxide/carbon-dioxide/nitrogen/water-vapour mixtures.

For each gas mixture studied and to which sulphur dioxide was added, a similar mixture free from sulphur dioxide was also studied for purposes of comparison. The results of all these experiments may be briefly summarized as follows.

Although some iron sulphide may penetrate the surface of white-iron bars annealed in atmospheres containing only dried carbon monoxide and sulphur dioxide, the depth of penetration of the sulphide is very slight and its distribution is quite different from

Table VIII
DEPTH OF PEELED LAYER IN TEST-BARS

Bar No.	Annealing Atmosphere	Depth of Peel	
		Maximum, in.	Minimum, in.
28	94% CO + 5% CO_2 + approx. 1% SO_2 ...	0·016	0·010
29	89% CO + 10% CO_2 + approx. 1% SO_2 ...	0·036	0·021

that responsible for peeling. The addition of carbon dioxide to carbon monoxide and carbon monoxide plus sulphur dioxide mixtures caused a greatly increased amount of decarburization to occur and a much deeper penetration of iron sulphide. The usual oxide phase followed the penetration of the sulphide, resulting in structures typical of the peeled layer. Increasing the proportion of carbon dioxide up to 20% in carbon-monoxide/carbon-dioxide mixtures increased the amount of decarburization, as would be expected. Further, increasing the proportion of carbon dioxide in carbon-monoxide/carbon-dioxide/sulphur-dioxide mixtures caused an increase in the amount of decarburization and an increase in the depth of penetration of iron sulphide and the oxide phase, forming a layer of peel. Figures 17 and 18 illustrate the appearance of the edge structures of bars 28 and 29 annealed for 100 hr. at 1000° C. in the gas mixtures given in Table VIII, and the effect of the higher carbon dioxide content in the gas mixture in causing an increase in the depth of the peel and the extent of decarburization can clearly be seen.

When nitrogen was introduced to replace part of the carbon monoxide in carbon-monoxide/carbon-dioxide mixtures, the experimental results showed a similar trend, that is, as the proportion of carbon dioxide in the mixtures increased there was a consequent increase in the amount of decarburization. The depth of the peeled layer formed in bars annealed in nitrogen/carbon-monoxide/carbon-dioxide/sulphur-dioxide atmospheres increased as the carbon dioxide content of the atmosphere increased.

When bars were annealed in atmospheres containing water vapour, decarburization occurred to a greater extent than in bars annealed in similar but dried atmospheres, and the higher the proportion of carbon dioxide present in the atmospheres containing water vapour, the greater was the amount of decarburization effected. When sulphur dioxide was introduced, the appearance and depth of penetration of iron sulphide and the oxide phase were typical of peeled samples. The depth of the peel was measured in a number of experiments and in each case it was found that deeper peel was produced in the bars annealed in atmospheres containing water vapour than in corresponding dry atmospheres. The deepest peel was observed in the bars which had been decarburized to the greatest degree. The figures obtained for the depth of the peeled layer in two samples are given in Table IX, and comparison of these figures with those given in Table VIII indicates the extent to which the depth of the peel is increased by the presence of water vapour in otherwise identical atmospheres. Figure 19 indicates the structure at the edge of bar 31, which was annealed for 100 hr. at 1000° C. in the atmosphere indicated in Table IX.

Table IX
DEPTH OF PEELED LAYER IN TEST-BARS ANNEALED IN ATMOSPHERES CONTAINING WATER VAPOUR

Bar No.	Annealing Atmosphere	Depth of Peel	
		Maximum, in.	Minimum, in.
30	94% CO + 5% CO_2 + approx. 1% SO_2 + water vapour ...	0·041	0·035
31	89% CO + 10% CO_2 + approx. 1% SO_2 + water vapour ...	0·060	0·049

In all gas mixtures containing carbon dioxide, increasing the amount of sulphur dioxide present caused an increase in the amount of iron sulphide absorbed by the metal to form a peeled layer containing a greater amount of iron sulphide.

The variations in carbon and sulphur from edge to centre were determined chemically for many of the bars annealed in gaseous mixtures, using the technique of machining away peripheral layers, as described previously. The results of these determinations will not be reported in full, and only the salient features are discussed in what follows.

The determination of total carbon in this manner showed that it decreased from the centre to the edge for atmospheres free from and also containing sulphur dioxide, in the manner normal for partially decarburized materials. The microstructures of commercial peeled castings usually show variations in carbon from edge to centre which would not be expected to correspond with the normal variations obtained when decarburization is not accompanied by sulphur penetration. For example, there is usually a segregation of divorced pearlite near the sulphide concentration, which may be followed, towards the centre of the section, by a completely ferritic band. In the case of the small samples referred to here, which were decarburized by gaseous mixtures, the final cooling rate was relatively rapid compared with that usual in commercial furnaces. The author finds that rapid cooling tends to eliminate the segregation of divorced pearlite which is usually characteristic of commercial samples. It is suggested that this peculiar distribution of carbon arises at some stage during the slow cooling and that the carbon distribution is normal at the top annealing temperature, that is, in the range 950–1050° C. Figure 20 shows the structure of a peeled sample which was air-cooled from 1000° C., and it can be seen that the carbon content of the sample increases progressively from the edge towards the centre.

Test-bars annealed in carbon monoxide alone, and in carbon-monoxide/carbon-dioxide/nitrogen atmospheres with and without water vapour showed a decreasing sulphur content from the centre towards the surface, indicating that desulphurization had

occurred. However, when sulphur dioxide was introduced into these atmospheres, penetration of sulphur into the metal was obtained, deeper penetration accompanying increasing carbon dioxide contents but constant sulphur dioxide contents in the annealing atmosphere. In all cases the highest sulphur contents occurred at the outer surface of the bars. From this surface inwards there was a gradual decrease in sulphur, until at the limit of the peel a sharp drop in sulphur content to that of the original white iron was observed. Plotting the sulphur content against the distance from the surface produces a curve similar to that of curve *S* in Fig. 29. It will be noted that this distribution of sulphur differs from that illustrated in Fig. 25, when bars were annealed in ore of high sulphur content. It is suggested that the reason for these differences in sulphur distribution through the peeled layer for samples annealed in ore on the one hand, and in a gaseous mixture on the other, is due to the fact that in the latter case the concentration of sulphur in the annealing atmosphere was maintained throughout the whole annealing cycle, whereas in the former case the liberation of sufficient sulphur into the annealing atmosphere to cause absorption of sulphur by the metal ceased after a certain time, and at a later stage some sulphur was removed from the surface of the bars by the annealing atmosphere.

SUMMARY AND GENERAL DISCUSSION

It is felt that the principal factors influencing the formation of peel in whiteheart malleable cast iron have been explored in the experiments described in this paper. It is obvious that the phenomenon of peeling is very complex and is associated with the removal of carbon by an oxidizing gaseous medium. This is accomplished at the surface of the casting and requires the diffusion of carbon from the centre and the simultaneous absorption of sulphur, as iron sulphide, by the reaction of sulphur dioxide in the annealing atmosphere with metal at the surface of the casting. This sulphide tends to diffuse into the casting as the carbon is removed. These two diffusion processes are most probably interrelated. Consequent upon the inward diffusion of sulphur there is also an inward diffusion of oxygen in the form of an oxide or silicate phase.

The experiments which have been carried out clearly indicate that peeling, as it is understood in the whiteheart malleable-iron industry, will not occur unless there is an external supply of sulphur during the annealing of the white iron. This sulphur may originate as sulphides or sulphates in the annealing ore, or as sulphur-containing gases arising from the combustion of the heating fuel. It is considered that this point is the most important from the practical aspect, and that when trouble is experienced attention should first be paid to the sulphur content of the ore. Various factors have been shown to inhibit or aggravate peeling when the external supply of sulphur is present, but resort should only be made to adjusting these secondary factors when the external supply of sulphur cannot be eliminated.

If the external supply of sulphur is present then the following variables influence whether peeling actually occurs, and, if it does occur, the severity of the peel :

(1) The actual sulphur content of the ore or the annealing atmosphere. For metal of a given composition, the higher this value for sulphur the greater will be the tendency to peel.

(2) The manganese and sulphur contents of the metal being annealed. The higher the ratio of the manganese content to the sulphur content the lower is the tendency to peel formation. For a given set of conditions it is possible, by suitably increasing the manganese content relative to the sulphur content, completely to inhibit peeling.

(3) The decarburizing power of the annealing medium. The more strongly decarburizing the annealing medium, the greater is the depth of peel. For ore annealing this implies that a high new-ore/used-ore ratio will encourage peeling for a given sulphur content. For purely gaseous decarburization, stronger decarburizing conditions are obtained with higher ratios of carbon dioxide to carbon monoxide. The influence of increasing the strength of annealing ore is probably the result of a consequent increase in the concentration of carbon dioxide in the atmosphere of the annealing can.

The effective decarburizing power of the annealing medium is also increased by the presence of water vapour in the annealing atmosphere so that the depth of peel becomes greater with increasing amounts of water vapour, all other factors remaining constant.

The sulphur content of a given ore mixture may increase with use by the absorption of sulphur from the casting and from the products of combustion of the heating fuel. Thus a used ore may have a higher sulphur content than corresponding new ore, in which case an increase in the ratio of new ore to used ore gives a lower average sulphur content in the aggregate and thus, in spite of its greater decarburizing power, there is a reduction in the tendency to peel formation.

(4) The period of time at the top annealing temperature. The longer the castings are held in contact with the sulphur-containing annealing medium at elevated temperatures (950–1050° C.) the deeper will be the resultant peeled layer.

(5) The actual value of the top annealing temperature. Up to 1050° C., at least, the higher the annealing temperature the greater is the depth of peel.

(6) The silicon content of the metal. All other factors remaining constant, the higher the silicon content the greater is the tendency to peel and the greater is the depth of peel.

(7) The initial carbon content of the metal. The lower the carbon content, the greater is the depth of the peel.

It should be possible, in the opinion of the present investigator, to explain the influence of these variables in a single comprehensive theory, but before this is attempted it is advisable to recall the metallurgical characteristics of the material being annealed and the annealing process.

The composition of whiteheart malleable cast iron usually falls within the following limits :

Total carbon	...	...	2·8 –3·7%
Silicon ...	...	...	0·3 –0·8%
Manganese	...	...	0·1 –0·5%
Sulphur ...	...	...	0·07–0·4%

In the as-cast condition the material usually has a typical hypo-eutectic white-iron structure consisting essentially of iron carbide (cementite) and pearlite. At the annealing temperature the pearlite becomes transformed to austenite, a solid solution of carbon in γ-iron, so that there exists an equilibrium mixture of austenite and iron carbide. Decarburization in atmospheres containing carbon dioxide takes place predominantly by a reaction at the surface of the casting, according to the equation :

$$C \text{ (in austenite)} + CO_2 \longrightarrow 2CO.$$

The extent to which this process can occur is governed principally by the diffusion of carbon in the austenite from the centre of the section to the surface of the casting. As carbon is removed from the austenite, the massive eutectic cementite gradually dissolves to make up this loss. As a result of these decarburization and diffusion processes, there is usually an increasing carbon content from the edge to the centre of the casting, in accordance with the laws of diffusion. In heavy sections when the time of annealing does not allow complete decarburization, the composition of the iron is adjusted so that any remaining cementite is decomposed to nodular graphite.

The mode of occurrence of the sulphur depends upon the manganese and sulphur contents. In the absence of manganese the sulphur exists as iron sulphide (FeS). Sulphur is soluble in—and hence can diffuse through—austenite, although carbon may reduce the extent of this solubility. Manganese tends to combine preferentially with sulphur according to the reaction :

$$FeS + Mn \rightleftharpoons MnS + Fe.$$

When the manganese content is in excess of about 1·7 × sulphur content, all the sulphur tends to be combined with the manganese as manganese sulphide. The reaction between iron sulphide and manganese is reversible, so that a slight excess of manganese may be required in order completely to neutralize the sulphur as manganese sulphide. Manganese sulphide is probably much less soluble in austenite than is iron sulphide. Any excess of manganese over that required to neutralize the sulphur is usually regarded as being either combined with the carbide as $(Mn, Fe)_3C$ or dissolved in the austenite. When the manganese content is insufficient to neutralize the sulphur, then both iron sulphide and manganese sulphide are seen in the microstructure. These compounds are appreciably intersoluble and so are never present in the pure forms. The solubility of iron sulphide in manganese sulphide increases markedly with increasing temperature.

When annealed in an oxidizing atmosphere, that is, containing carbon dioxide, which is free from sulphur, desulphurization can occur. Annealing in ore of low sulphur content usually has a desulphurizing effect.

The silicon content of a white iron is usually considered to be dissolved in the austenite, although some may be combined with the cementite. The small amount of phosphorus present in the metal is usually dissolved in the austenite.

Thus at a relatively early stage in the annealing of whiteheart malleable cast iron there exists an outer layer of austenite, of zero carbon content at the extreme edge, increasing to the limit of solid solubility towards the centre ; this austenite also contains silicon and may also contain manganese and some phosphorus in solid solution.

When pure iron is annealed in contact with an external supply of sulphur, penetration of sulphur occurs, and this process has been studied by Bramley and his co-workers,[6] but Desch[7] has pointed out that this penetration does not follow the simple law of diffusion. Bramley's results have been considered more recently by Dorn and Harder[8] and shown to fit a theory of interstitial diffusion. Whether the diffusion of sulphur into pure iron is a simple case of interstitial diffusion is not known. It was shown by Bramley and his co-workers that small amounts of non-metallic elements in solid solution in iron appreciably alter the diffusion constant for other non-metallic elements, and in the study of a series of steels with up to 0·965% of carbon they claim to have shown that carbon greatly retards the diffusion of sulphur. It should be pointed out that the steels used by these investigators contained up to 0·4% of manganese, which probably also influenced their results. It would be expected that the presence of excess manganese would tend to inhibit the diffusion of sulphur, owing to the formation of manganese sulphide, although the extent to which this can occur will also depend upon the ability of manganese to diffuse to meet the incoming sulphur. With a high external concentration of sulphur, it is conceivable that manganese sulphide could form in the initial stages, but with a steady supply of sulphur the manganese would become exhausted and the sulphur would form iron sulphide, which would dissolve the manganese sulphide. If the diffusion of sulphur in pure iron occurs interstitially, the maximum sulphur content obtainable would depend upon the solid solubility of sulphur in austenite at the temperature of the experiments. As the solubility of sulphur in pure iron decreases with decreasing temperature, much of this sulphur would be expected to be deposited from solid solution on cooling. From the cellular or grain-boundary distribution of the sulphide in samples of sulphurized pure iron this would appear to be the case, although this structure may be related to the formation of liquid sulphide complexes which will be referred to later.

To return to a consideration of the annealing of white iron of low manganese content in contact with an external supply of sulphur, penetration of sulphur by interstitial diffusion would not be expected to occur until an appreciable amount of carbon (which is also interstitially dissolved in austenite) has been removed by decarburization. This is borne out by the experimental results which have been presented in this paper. Annealing in non-decarburizing atmospheres (CO) containing sulphur did not give a

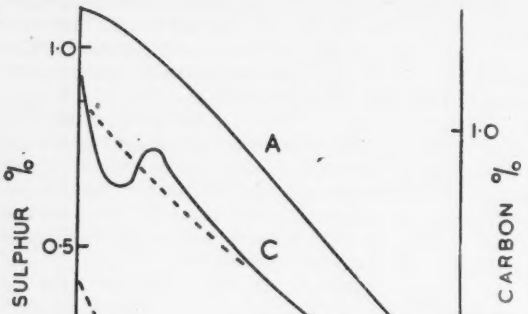

Fig. 28—Distribution of carbon and sulphur in steel after sulphurization followed by carburization (Bramley and co-workers)

deep penetration of sulphur, but as the decarburizing power of the atmosphere was raised by increasing the carbon dioxide and/or the water vapour content of the atmosphere, and, in the case of annealing in ore, when the ore strength was increased, there was a proportionate increase in the depth of the penetration of the sulphur. That the extent of decarburization is the primary influence in determining the depth of penetration of the sulphur was confirmed by observations on the depth of sulphide penetration when white-iron samples with various values for the ratio surface-area/mass-of-metal were annealed together in sulphur-containing ore. For a given set of decarburizing conditions, the amount of decarburization depends on the value of this ratio. Decarburization is a surface reaction, so that the greater the surface area relative to the mass, the greater is the decarburization. Thus a round bar will be decarburized less than a flat rectangular bar of the same cross-sectional area. It has actually been observed that the depth of sulphide penetration, and hence the thickness of the peeled layer, is greater the greater the relative surface area.

The concentration of sulphur which will dissolve in austenite is probably dependent upon the carbon content, as has been pointed out, and it appears that a certain minimum amount of carbon has to be reached by decarburization before sulphur penetration can begin in the manner which characterizes the peeling defect. This would explain why lower initial carbon contents and higher annealing temperatures cause increased depths of sulphur penetration—they are factors tending to give the necessary low carbon content more quickly, first at the surface and then to a greater depth.

The penetration of sulphur into white iron during decarburization provides a very interesting problem in the diffusion of two elements in solid solution, since the directions of diffusion are opposite. Bramley and his co-workers have studied the distribution of carbon and sulphur in carburized irons which had previously been sulphurized. These experiments were not the exact parallel of those conditions existing for peel formation, but some interesting comparisons are possible. A typical example of Bramley's results is illustrated in Fig. 28, where curve *A* shows the distribution of carbon in a sample of Swedish iron after carburizing for 10 hr. at 1000° C. The broken curve *S* shows the sulphur distribution after sulphurizing a similar sample for 40 hr. at 1150° C. The continuous curve *S* shows the distribution of sulphur after subsequent carburization for 20 hr. at 1000° C. The continuous curve *C* shows the distribution of carbon after this treatment. The interesting point of this experiment is the minimum on the carbon-concentration curve. It is possible that some desulphurization occurred during carburization, and this would explain the maximum on the final sulphur-distribution curve. It will be observed that the minimum on the carbon-distribution curve corresponds with the maximum on the sulphur-distribution curve. Bramley and his co-workers suggested that the original carbon distribution was of the type indicated by the broken line *C* and that on cooling, owing to the fact that the solubility of carbon in iron is reduced by the presence of sulphur, the carbon moved away from the region of high sulphur content on both sides. This explanation would seem feasible and is in line with the results obtained in the work reported here. When the external supply of sulphur was constant as in the gaseous experiments described in this paper, and when the samples were fairly rapidly cooled, the distribution of carbon and sulphur was of the type indicated in Fig. 29.

It should be noted that the sharp drop in sulphur content (curve *S*, Fig. 29) in rapidly cooled samples was not accompanied by any sharp change in carbon content (curve *C*). In slowly cooled samples annealed in ore containing sulphur, the distribution of carbon

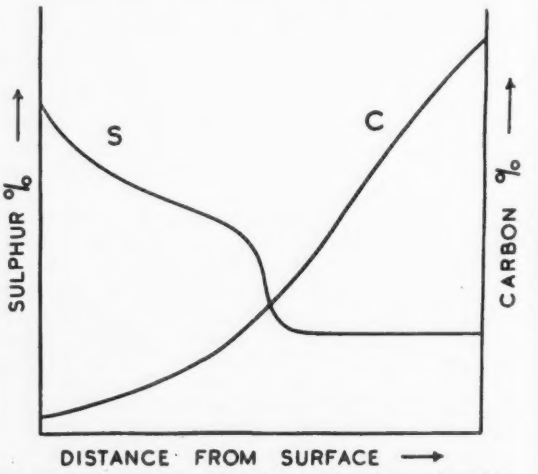

Fig. 29—Distribution of carbon and sulphur in white iron under conditions of constant external sulphur supply

and sulphur as determined by chemical analysis took the form shown in Fig. 25. It is suggested that this sulphur-distribution curve possesses a peak, owing to the fact that the external supply of sulphur became exhausted and desulphurization occurred during the latter part of the anneal in a manner similar to that experienced by Bramley during carburization. In all the rapidly cooled samples which the present investigator has examined, the carbon contents appear to vary uniformly from edge to centre without any sharp change at the limit of sulphur penetration. This is especially so from the point of view of metallographic examination, although the form of the pearlite appears abnormal in the area of the sulphur concentration. The apparently smooth gradation in carbon content from edge to centre in an air-cooled sample is illustrated in Fig. 20. This sample was annealed for 75 hr. at 1000° C. in ore containing 0·7% of sulphur.

Metallographic examination of slowly cooled peeled samples (Fig. 2) seems to indicate that the carbon distribution should take on the form given in Fig. 30 (curve *C*).

Chemical analysis of the peripheral layers has so far not demonstrated this form of carbon distribution, but this may be due to the relatively small scale of the structural layers and the resulting difficulty of obtaining representative samples for chemical analysis. It is possible, therefore, that the evidence of the metallographic examination gives a truer picture of the carbon distribution in slowly cooled samples. The usual curve for sulphur distribution on samples annealed in ore and slowly cooled is that given in curve *S*, Fig. 30. It is important to observe that the region of high sulphur content does not in this case correspond with a minimum in carbon content, but rather to a maximum point (that of the largest amount of the divorced pearlite). If Bramley's explanation of his curves was correct one might expect a minimum carbon content to occur coincident with the sulphur maximum in peeled samples. However, in the case of peeling it is possible that the band of pearlite in the peeled layer is related to the penetration of oxygen.

It will be seen that the characteristic carbon distribution across a peeled section shown metallographically in Fig. 2 and diagrammatically in curve *C*, Fig. 30, is the result of changes occurring at relatively low temperatures during slow cooling.

The divorced pearlite existing in the peeled layer may owe its presence also to the oxide phase. Oxides at relatively low temperatures have been observed by the present investigator and by Leroyer[9] to be responsible for the formation of granular carbide and divorced pearlite. A similar but much shallower layer of divorced pearlte and granular carbide occurs in normal whiteheart malleable cast iron and occasionally in blackheart malleable iron. Leroyer suggests that the more massive carbide particles are formed by the reaction :

$$3FeO + 5CO \rightleftharpoons Fe_3C + 4CO_2.$$

In the case of peeled samples this would involve diffusion of carbon monoxide and dioxide through the metal, which is considered highly improbable.

The concentration of sulphur in the peeled layer and particularly at the boundary between the peeled layer and the sound core frequently appears to be in excess of the limit of solubility of sulphur in solid solution. If this is so, some interest attaches to the reason for this high concentration being possible. The annealing temperature used in commercial annealing practice usually falls within the range 950–1050° C. The iron–sulphur system shows a eutectic of iron and iron sulphide (FeS) melting at 985° C. and having 30·8% of sulphur. It is known that the annealing temperature for malleable iron is frequently higher than 985° C., and it would appear possible for the liquid eutectic to be formed under sulphurizing conditions during annealing. If this is the case, then it would explain the occurrence of iron sulphide in excess of the limit of solid solubility and would allow of rates of diffusion of sulphur in excess of those permissible by interstitial diffusion through the solid.

So far, in considering the inward diffusion of sulphur, the influence of manganese has been disregarded. In actual fact it would only become an obstacle to the inward diffusion of sulphur when present in excess of that required to neutralize the initial sulphur content of the metal. When the manganese content is below this amount it would become gradually dissolved in any

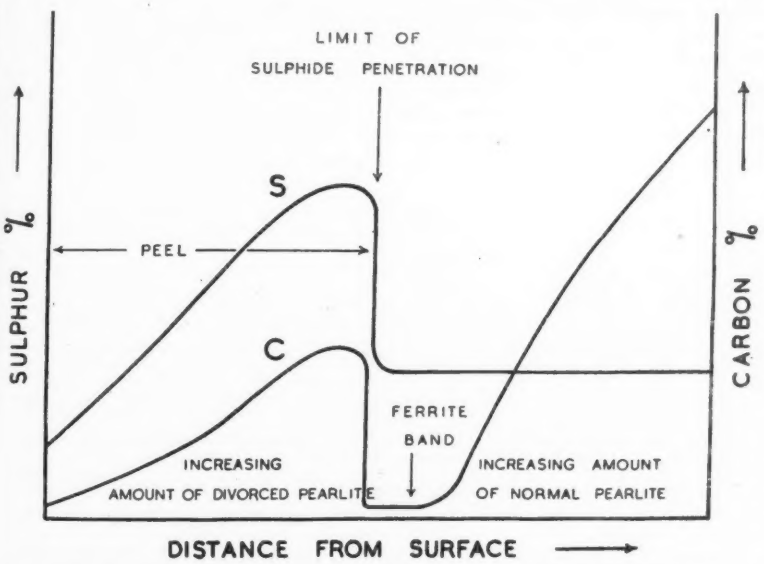

Fig. 30—Probable distribution of carbon and sulphur in commercial peeled casting

microscopically visible iron sulphide and perhaps precipitated again as manganese sulphide on cooling. When excess manganese is present, incoming sulphur would combine preferentially with the manganese to give manganese sulphide which, by virtue of its low solubility and high melting point, would not diffuse inwards. It is possible that sulphur can diffuse through the iron only by the interstitial diffusion of the elemental atoms. When the external supply of sulphur is continuously large during the anneal in relation to the internal supply of manganese, all the manganese will be used up in a given locality and the incoming sulphur will then diffuse normally. Thus manganese can only be considered as an inhibitor of peeling, reducing its severity, and not as a certain prevention of peeling. Whether an excess of manganese in the metal is successful in preventing peeling will depend, among other factors, upon the concentration of the external supply of sulphur, and the larger this is the less efficacious will be the manganese in the metal.

In describing the structure of the peeled layers, frequent reference has been made to the occurrence of an " oxide " phase in large amounts near the outer surface and gradually decreasing in amount on approaching the boundary of the peel. Furthermore, it has been shown that this phase occurs in intimate association with the sulphide phase. It is suggested in what follows that the occurrence of this phase is consequent upon the presence of, and is not the cause of, the high sulphur contents of the peeled layers. Metallographically, it is allotriomorphic, dark grey, and partially translucent. Examination under polarized light does not reveal it to be anisotropic. No positive identification has been attempted by the present investigator, but it would appear to be a silicate rich in iron and probably also containing some manganese. It is well known that the presence of sulphur in iron greatly increases its susceptibility to penetration by oxygen when annealed under oxidizing conditions. In the case of pure iron–sulphur alloys the FeS–FeO eutectic (melting point 940° C.) may be formed if the temperature is sufficiently high. It is suggested that in the case of peeling the iron sulphide forms first and then the atmosphere being oxidizing relative to this compound gives iron oxide, perhaps according to an equation of the type :

$$FeS + 3CO_2 \longrightarrow FeO + SO_2 + 3CO.$$

Similarly, silicon and manganese may also be oxidized, and the resultant oxides combine to form a silicate phase. Iron sulphide forms low-melting-point complexes with silicates, and it is possible that these could be formed in the liquid condition at the annealing temperature of whiteheart malleable iron. Since the oxide or silicate phase forms after the sulphide, it would be expected to be present in largest amounts at the outer surface of the casting, and this is in accord with observation. Where it is present in very large amounts, the sulphur may remain dissolved in the silicate, and this seems highly probable since, even when the concentration of sulphur is high at the outer edge of the peeled layer, if there is a large amount of silicate only a small amount of sulphide is microscopically visible. Further to exaggerate this effect, it is possible that oxidation of this type also proceeds during slow cooling at relatively low temperatures when decarburization has virtually ceased, owing to the drop in diffusibility of carbon, and sulphurization is prevented for the same reason.

The influence of silicon in promoting the formation of a peeled layer even in the presence of manganese is more difficult to explain with certainty. It is known that steels tend to become more susceptible to attack by sulphur-containing gases with increasing silicon contents. Careful examination of peeled samples containing manganese contents as high as 1% and also high silicon contents (up to 1·7%) has not revealed manganese sulphide in sulphide concentrations. This suggests that silicon is effective, directly or indirectly, in neutralizing the manganese and so preventing its combination with sulphur to give manganese sulphide. This could be accomplished in one or all of the following ways :

(*a*) The silicon could (by some reaction as yet unknown to the author) cause the preferential oxidation of the manganese.

(*b*) The silicon could enter into some form of chemical combination with the manganese. Manganese silicides are known, but not as microscopically visible phases in cast irons.

(*c*) The preferential volatilization of manganese from the solid metal, this process being encouraged by the silicon content.

No evidence is available in support of methods (*a*) and (*c*), but method (*b*) would be expected to influence the reaction between manganese and sulphur in cast iron, and Morrogh[10] has previously suggested, in an entirely different context, that silicon may influence the manganese and sulphur reaction.

The effect of silicon in promoting graphitization is considered to be a very unlikely reason for its effect on peel formation, as rapid graphitization tends to reduce the extent of decarburization.

Acknowledgments

The author desires to thank the Council and Director of the British Cast Iron Research Association for permission to publish this paper. To Mr. H. Morrogh the author tenders his sincere thanks for his helpful advice during the experimental work and in the checking of the manuscript.

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The Creep Properties of Molybdenum, Chromium–Molybdenum, and Molybdenum–Vanadium Steels

By J. Glen, B.Sc., A.R.T.C., A.I.M.

SYNOPSIS

The numerous variables to be considered in the study of creep are discussed at length, and the importance of controlling, and where possible, of isolating these variables is emphasized. Attention is drawn to the limitations of short-time tests, particularly when estimating the creep characteristics of a new type of steel. Creep tests at different temperatures and stresses have been carried out on 0·5%-molybdenum, 0·8%-chromium/0·5%-molybdenum, and 0·5%-molybdenum/0·25%-vanadium steels, and the effect of carbon, silicon, manganese, and aluminium have been determined. The effects of various heat-treatments have also been examined. Rupture tests of up to 58,000 hr. duration have been carried out and the mode of failure discussed. By an extensive series of long-time creep tests, the stress/temperature relationships for 0·1% and 0·2% deformation in 100,000 hr. have been estimated for the molybdenum–vanadium steel and suitable short-time tests for checking the quality of such steels are discussed.

The results show that the molybdenum–vanadium steel in the normalized and tempered condition has creep and rupture properties much superior to those of the molybdenum and chromium–molybdenum steels.

INTRODUCTION

THE history of creep testing may be roughly divided into three periods. When it was first realized that metals under constant load at high temperature continued to deform over indefinite periods of time, the methods devised for studying the problem were somewhat crude. Although very useful pioneer work was done, there arose the erroneous assumption that at any temperature there was a limiting creep stress at and below which deformation ceased after a short time. As refinements were made to the equipment used, the limiting creep stress assumed progressively lower values until it was realized that the limit was set by the accuracy of the extensometer used and by the duration of the test.

The next period was that in which creep rate became all important. It was found that after the initial period of a creep test had been completed at the testing temperature, the steel appeared to settle down to a uniform creep rate. Thus, on the basis of tests lasting say 1000 hr., it is possible to obtain a stress/temperature relationship for any specified deformation. Neglecting the deformation occurring in the initial period of the creep tests and assuming the creep rate to remain constant over longer periods of time, then a creep rate of 10^{-8} in./in./hr. measured at 1000 hr. is equivalent to a deformation of 0·01% in 10,000 hr. or 0·1% in 100,000 hr. (Fig. 1*a*, *b*, and *c*).

In the United States of America this appears to be the recognized method of presenting creep data.[1, 2] As will be made clear later this method is not reliable, for below a certain temperature, depending upon the steel and not ascertainable by the method of testing, the estimated deformation becomes progressively too low, whilst above that temperature it becomes increasingly an overestimate of the actual behaviour of the steel.

The third and present period of creep testing was initiated by Bailey and Roberts[3] as long ago as 1932, who stated the problem succinctly as follows:

"The simple demand upon any material is that it shall endure the normal conditions of service without failure. This requires that the material shall *not*:

(*a*) Deform more than working conditions permit.

(*b*) Undergo change in condition which may endanger the serviceability intended."

Changes in the condition of the steel arising from prolonged exposure at high temperatures are: deterioration of creep resistance due to spheroidizing, scaling in air or steam, and embrittlement. Scaling and embrittlement are discussed later; deterioration of creep resistance must obviously be considered in relation to the first requirement stated above.

Considering this requirement, it is now generally agreed that with alloy steels for high temperature service the permissible deformation should be limited to small amounts in order to obviate any danger of intercrystalline failure which, however, under ordinary conditions is preceded by more than 1% of deformation. A second reason for this limitation is that even in the absence of intercrystalline cracking, failure may occur in a time not much greater than that for a comparatively small deformation. Thus, working-stress data for alloy steels are usually presented as the stress/temperature relationship to give a deformation of 0·1–0·5% in a specified time. A suitable maximum time is 100,000 hr. ($11\frac{1}{4}$ years).

The compilation of such data can at present be achieved only by carrying out a large number of long-time tests and by employing suitable extrapolation methods to estimate therefrom the stress for any given deformation in any given time. The published information of this type is rather meagre. Bailey, who was the first to carry out such tests, investigated one carbon and one molybdenum steel[4]. Tapsell and others have investigated three carbon[5] and three molybdenum steels[6] taken from service components, and a molybdenum steel casting.[7] A summary of the results is shown in Fig. 2 (*a* and *b*).

It will be noted that in all of these curves there is

Manuscript received 21st June, 1947.

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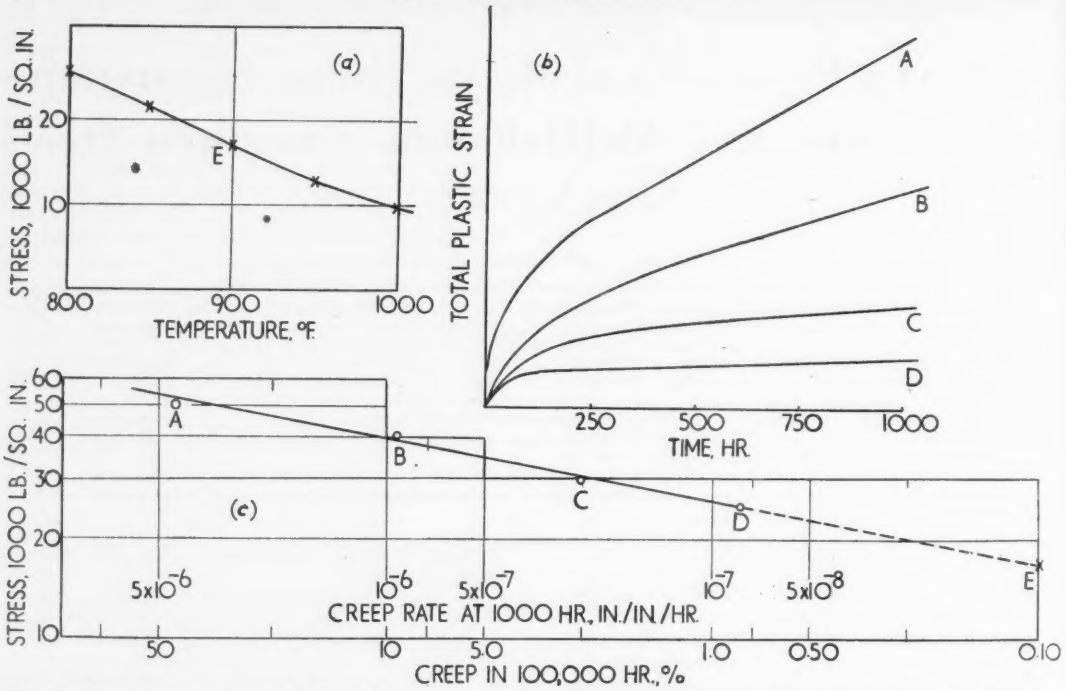
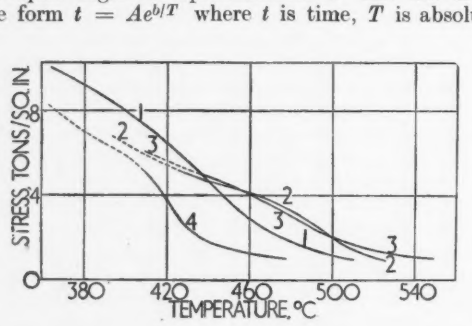


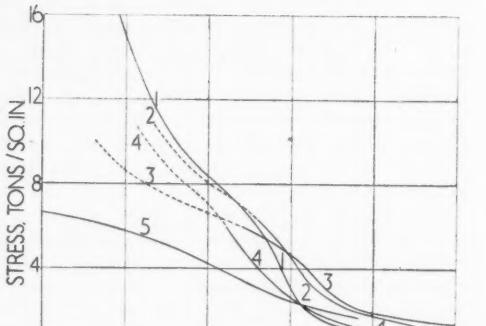
Fig. 1—Stress/temperature relationship for 0·1% of plastic strain, derived from the creep rate at 1000 hr. (*a*) Stress/temperature curve for 0·1% of creep in 100,000 hr. derived from (*c*) and similar tests at other temperatures. (*b*) Creep tests at 900° F. and at various stresses. (*c*) Log.-stress/log.-creep-rate curve derived from (*b*)

a bulge followed by a sharp drop in working-stress values over a narrow range of temperature. An explanation of this effect was given by Bailey and Roberts[3, 4] who, as a result of extensive investigations on carbon and molybdenum steels, found a quantitative relationship between time and temperature for an equal degree of spheroidization. The law is of the form $t = Ae^{b/T}$ where t is time, T is absolute



Curve	Reference	Source	Deoxidation	Carbon, %
1	Bailey[4]	Forged steel	—	0·4
2	Tapsell[5]	Superheater header	Si-killed	0·24
3	Tapsell[5]	Steam-pipe	FeMn to bath, Al and Alisifer to ladle	0·135
4	Tapsell[5]	Superheater tube	2 lb./ton of Al to ladle	0·12

Fig. 2 (*a*)—Stress/temperature curves of carbon steels for 0·1% of plastic strain in 100,000 hr.



Curve	Reference	Source	Treatment	Carbon, %
1	Bailey[4]	5-in. dia. bar	W.Q. 860° C., T. at 650° C. for 1 hr.	0·29
2	Tapsell[6]	Superheater header	N. 920° C., T. at 660–670° C.	0·26
3	Tapsell[6]	Steam-pipe	N. 920° C.	0·09
4	Tapsell[6]	Superheater tube	N. 940° C. plus open anneal at 700° C.	0·17
5	Tapsell[7]	Casting	18 hr. at 900° C. Furnace cooled to 210° C. in 30 hr., plus 3 hr. at 650° C. furnace cooled to 225° C. in 15 hr.	0·31

Fig. 2 (*b*)—Stress/temperature curves of 0·5% molybdenum steels, for 0·1% of plastic strain in 100,000 hr.

temperature, and A and b are constants. On the assumption that changes in creep depended solely upon spheroidization it was concluded that the effect of prolonged exposure to relatively low temperatures could be determined by short-time exposures at higher temperatures. It was demonstrated by short-time creep tests on normalized steel at a certain temperature that a short tempering at a higher temperature before testing improved the creep strength, whereas longer tempering times caused increasing deterioration. An example of this effect is shown in Fig. 3. Thus in long-time creep tests in the normalized condition, up to a certain time the creep rate will fall due to the initial stiffening effect of the testing temperature, and thereafter the creep rate will increase due to subsequent softening. This effect explains the bulge and subsequent sharp drop in the stress/temperature curves shown in Fig. 2 (*a* and *b*), and is sufficient to condemn extrapolation methods based on creep rate alone, as shown in Fig. 1 (*c*). As is demonstrated below, the effect of temperature is more complex than was envisaged by Bailey.

The important variables to be considered in the study of creep are :

Test Conditions:

(*a*) Temperature
(*b*) Stress
(*c*) Time
(*d*) Atmosphere

Type of Steel:

(*a*) Composition
(*b*) Method of manufacture and deoxidation
(*c*) Fabrication
(*d*) Heat-treatment and microstructure

TEST CONDITIONS

In service, steel is subjected to combined stresses, but most high-temperature work has been carried out in the form of tensile tests under constant stress, and for the present purpose it is sufficient to restrict attention to this type of test.

Temperature, Stress, and Time

The results of a series of typical creep tests at constant temperatures and at different stresses, on two steels A and B, are shown in Fig. 4. At the higher stresses steel A is superior at all stages of the creep test and takes the longer time to fracture. At a lower stress steel B is better for a short period of time, and thereafter steel A is superior. As the stress is further decreased steel B is the better over an increasingly long period of time, and a stage may be reached when it is superior even if the tests are continued until fracture.

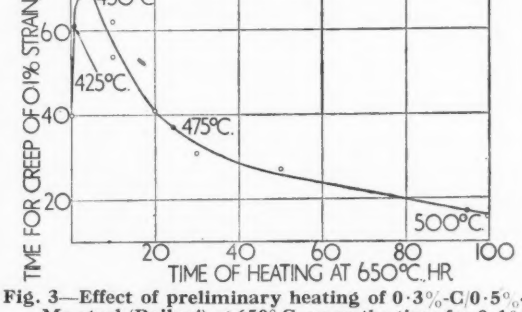


Fig. 3—Effect of preliminary heating of 0·3%-C/0·5%-Mo steel (Bailey[4]) at 650° C. upon the time for 0·1% of plastic strain. Temperature, 560° C.: Stress, 6 tons/sq. in. Black dots show calculated effect of 100,000 hr. at temperature indicated

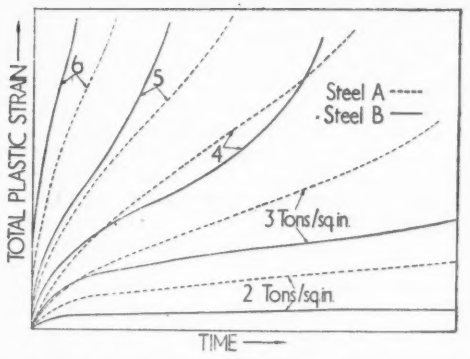


Fig. 4—Typical creep curves of two different steels A and B at different stresses and constant temperature

The relationships shown in Fig. 4 may vary significantly when the tests are carried out at higher or lower temperatures. Accordingly, when comparing the creep properties of steels temperature, stress, and time should be considered together, since the effect of any one of them depends upon the conditions in respect to the other two.

In creep testing, it is the load and not the stress which is kept constant, so that as the test-piece extends the true stress increases due to the reduction of area. This effect is negligible when the deformation is small but is of great importance in rupture tests.

Atmosphere

Creep tests or rupture tests at high temperature where scaling and/or intercrystalline cracking occurs are also affected by a variation in the true stress but for a different reason. In this case, the results are influenced by the size of the test-piece used. The smaller the test-piece the greater the percentage of reduction in area due to scaling in a given time, and the greater the influence of an intercrystalline crack forming on the surface of the test-piece. The more severe the effect of the atmosphere as regards scaling or intercrystalline cracking, the greater is the effect of test-piece size. It must, however, be remembered, that even if the test-piece size is standardized, the scaling effect is of increasing importance as the temperature rises and the time of test increases, and it will therefore have an influence on the accuracy of extrapolation methods for obtaining the long-time creep properties of the steel.

TYPE OF STEEL

Composition

Since only a few carbon steels and a few 0·5% molybdenum steels have been tested for long-time creep properties (Fig. 2*a* and *b*), the effect of composition has been studied mainly on a basis of short-time creep tests, most of which have unfortunately been inadequate to give a true picture of the effect

of composition. For example, even in the simple case of carbon steels, a review of the literature shows no real agreement on the effect of carbon. Depending on the test used, increasing the carbon content is said to be good, bad, or indifferent at any temperature up to 1000° F. It is not surprising, therefore, that little work has been published on the effects of minor differences in composition. On the existing evidence, however, it appears that such differences, even within specification limits, may have a considerable influence on the creep properties in short-time tests.

Method of Manufacture and Deoxidation

The method of manufacture has an important bearing on the creep properties of steel, and various processes have been credited with producing steels with the best creep resistance, again on the basis of short-time tests. In the author's opinion, however, there is no ' better steel ' associated with a particular process of manufacture *per se*. The quality is directly related to the degree of deoxidation of the steel and the effect of this on the structure obtained on heat-treatment. For example, low carbon open-hearth steels are usually more highly oxidized than similar electric-furnace steels, and in general, for the same normalizing temperature the structure is coarser. Thus, comparing an open-hearth steel with an electric-furnace steel, normalized at the same temperature, is equivalent to comparing one steel in different conditions obtained by normalizing from different temperatures. If this structural effect is taken into consideration, the influence of the process of manufacture is placed in its proper perspective. In general, electric-furnace steel requires less aluminium to give a fine grain-size (McQuaid-Ehn grain-size 6–8). In the fine-grained condition there is nothing to choose between open-hearth steel and electric-furnace steel.

Fabrication

Fabrication processes break down the ingot structure which may then be coarse, fine, or even slightly cold-worked, depending on the finishing temperature. Subsequent normalizing tends to remove such effects, but a coarse structure tends to remain coarse. Where the finished article is not heat-treated as a whole, the effect of welding, corrugating, bending, etc., must be taken into consideration, since localized heating implies that some part of the steel has been heated to any temperature from atmospheric to that of the localized treatment. Cold working is to be avoided, since among other things it tends to promote spheroidizing, with resulting poor creep properties. The surface of the steel is also affected by the fabrication processes, and incipient burning or intercrystalline penetration of oxide may be present in varying degrees. Such effects will tend to initiate intercrystalline cracking.

Heat-Treatment and Microstructure

The heat-treatment usually specified for the steels under consideration is normalizing. Increasing the normalizing temperature tends to coarsen the final structure and may also promote the solution of sluggish carbides. Very considerable differences in creep properties may thus result from different normalizing treatments. After normalizing, the steel may be tempered, or after working it may be given a stress-relieving treatment. Such treatments can also have great effects on the creep properties.

Furnace cooling from the normalizing temperature, *i.e.*, annealing, is also used, particularly with steels in the cast condition. Due to the slow rate of cooling through the tempering range, this treatment will have different effects depending on the composition of the steel. Furnace cooling is detrimental to creep resistance.

It is clear, therefore, that steel may be put into service with widely different microstructures due to the cumulative effects of the above variables though, as a rule, little cognizance is taken of them.

If these variables acted independently it would be a relatively simple matter to study each in turn. The fact that some of them are closely interrelated increases the complexity of the subject and necessitates a more detailed study than has hitherto been carried out.

Since it is virtually impossible to test completely even one steel in one condition of heat-treatment it is necessary to use comparative tests, and indeed the literature on creep may be said to consist almost solely of such tests. These, however, must refer to

Table I

MANUFACTURING DETAILS AND ANALYSIS OF THE MOLYBDENUM STEELS

Steel	Process	Method of Deoxidation	Analysis							
			C, %	Si, %	S, %	P, %	Mn, %	Ni, %	Cr, %	Mo, %	Cu, %
A1	18 lb. high-frequency furnace	FeMn and FeSi	0·055	0·16	0·032	0·018	0·48	0·10	0·04	0·50	0·08
A2	Basic open-hearth	SiMn to ladle	0·09	0·22	0·044	0·013	0·43	0·09	0·03	0·51	0·08
A3	,, ,,	SiMn to ladle	0·145	0·15	0·030	0·032	0·56	0·12	0·01	0·48	0·06
A4	,, ,,	SiMn to ladle	0·21	0·17	0·030	0·029	0·49	...	...	0·55	...
A5	Acid open-hearth	SiMn to ladle	0·31	0·18	0·038	0·040	0·55	...	...	0·52	...
A6	18 lb. high-frequency furnace	FeMn and FeSi	0·395	0·14	0·029	0·018	0·48	0·11	0·06	0·48	0·08
A7	,, ,, ,,	FeMn and FeSi	0·10	0·16	0·030	0·019	0·52	0·10	0·05	0·52	0·08
A8	600 lb. high-frequency furnace	FeMn and FeSi	0·12	0·10	0·025	0·018	0·53	...	...	1·33	...
A9	,, ,, ,,	FeMn and FeSi	0·13	0·06	0·017	0·015	0·44	...	...	0·55	...

one or more of three objects: To check whether a particular batch of steel is up to standard; to find the optimum heat-treatment conditions for a given steel; or to examine the effect of changes in composition. It is essential that such tests indicate whether a particular steel is better than or inferior to a steel whose long-time creep properties are known.

Johnston and Tapsell[8] examined a large variety of the better known short-time creep tests on carbon steels and found that none of them could be relied upon to give an estimate of long-time creep properties, different tests placing the steels in a different order of creep resistance. For these same tests even wider variations in creep resistance would be found with alloy steels.

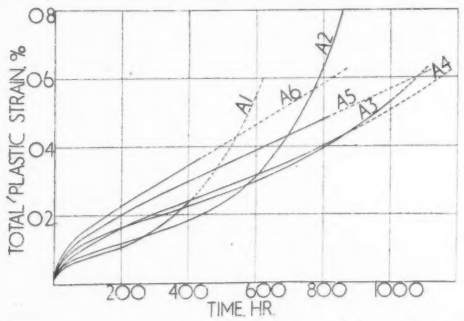
In view of the importance of short-time creep tests, it is remarkable than no attempt appears to have been made to reconcile the conflicting results obtained and to place short-time creep testing on a rational basis.

Any particular criterion of creep resistance based on a short-time test may be taken as being representative of one point on a particular creep curve, *i.e.*, a definite amount of deformation after a certain time at a particular stress and temperature. Conflicting results arise, since the various criteria represent either different points on the same curve or points on different curves. In other words sufficient cognizance has not been taken of the many variables.

The minimum to which the problem can be simplified is by the study of one steel variable at a time, *e.g.*, the effect of varying the carbon content or of different heat-treatments, but at the same time the four testing variables deformation, time, stress, and temperature must also be examined.

Since the problem at its simplest is one of six dimensions, it is clear that the steel variables must be held under very strict control otherwise wide scatter will be obtained in the results.

In the work described below, the method of approach has been to plot a steel variable against the time to reach specific amounts of deformation at a constant temperature and stress. It was impracticable to repeat the tests with a wide variety of stresses and temperatures, but by employing temperatures of 550° and 600° C. and stresses of 4, 6 and 9 tons/sq. in. a clear indication of the effect of these variables was obtained. These temperatures and stresses were chosen because a number of tests had already been carried out under these conditions, and from experience it was known that the tests would be neither too short nor too long.



Steel ...	A1	A2	A3	A4	A5	A6
Carbon Content, %	0·055	0·09	0·145	0·21	0·31	0·395

Fig. 5 (*a*)—Curves of creep tests on 0·5% Mo steels with various carbon contents. Stress, 4 tons/sq. in.: Temperature, 600° C.

Three types of steel have been used in this investigation: 0·5% molybdenum steel, 0·5%-molybdenum/0·8%-chromium steel, and 0·5%-molybdenum/0·25%-vanadium steel. Apart from these variations, the composition variable has been studied mainly in respect to carbon and silicon. The effect of varying the heat-treatment has also been examined in some detail.

More prolonged tests were carried out to estimate the rupture strength in 100,000 hr. The stress/temperature relationship for 0·1% of deformation in 100,000 hr. has also been estimated for a molybdenum–vanadium steel in the normalized condition and in the normalized and tempered condition.

CREEP TESTS ON 0·5% MOLYBDENUM STEEL

The manufacturing details and complete analyses of the steels are listed in Table I. The carbon content of the steels varied from 0·05 to 0·4%, the other elements being essentially constant. With the exception of the 0·05% and the 0·4% carbon steels, which were made in an 18 lb. high-frequency furnace, all of the steels were from open-hearth production casts.

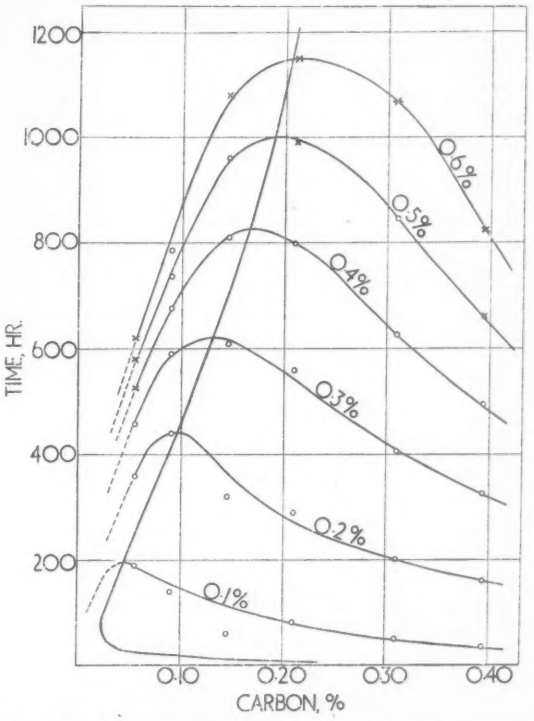


Fig. 5 (*b*)—Curves showing the effect of carbon content of 0·5% Mo steel on the time to reach the plastic strains indicated. Derived from Fig. 5 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

The steels, in the form of 1-in. dia. bar, were normalized from 950° C., with the exception of the 0·05% carbon steel which was normalized from 1000° C. to give a structure similar to that of the 0·09% carbon open-hearth steel. The object of these treatments was to test the steels in the condition which would give maximum creep resistance in long-time service at high temperature.

EFFECT OF CARBON CONTENT

The results of creep tests at a temperature of 600° C. and a stress of 4 tons/sq. in. are shown in Fig. 5 (*a*). Comparing the time to reach the point 0·1% of plastic strain the steels, with one exception, fall in order of carbon content, the lowest carbon steel taking the longest time. At a greater deformation a higher carbon steel takes the longest time. By plotting the carbon content against time to reach a specific amount of deformation and by drawing a family of curves as shown in Fig. 5 (*b*) a much clearer picture of the effect of carbon is obtained.

For any specific amount of deformation a maximum time is obtained at a definite carbon content and, as shown, a smooth curve can be drawn through these maxima. The form of this curve (described hereafter as the 'maxima' curve) can be explained as follows.

As the carbon content increases there is an increase in the elastic limit and a decrease in the plastic strain on loading; also the slower is the rate of creep in the first few hours of the test. Thus, with a very small deformation (less than 0·01%) the maximum time is obtained with a carbon content greater than 0·2%. With increasing deformation up to 0·05%, the maximum is obtained at progressively lower carbon contents, due to the greater latent resistance to creep of these steels (*see* p. 44).

As shown in Fig. 5 (*a*), at about 0·1% of deformation the steel with the lowest carbon content enters the third phase of creep, *i.e.*, the creep rate begins to increase with time, whereas with higher carbon contents this occurs with larger amounts of deformation. Thus, below about 0·05% of deformation and at about 0·025% of carbon, the maxima curve changes direction and shifts to higher carbon contents as the deformation increases. The curve asymptotically tends to reach a carbon content of approximately 0·25%. Unless other effects intervene the 0·25% carbon steel will give a maximum time to rupture under these conditions of test.

Further tests were carried out at 550° C. and a stress of 9 tons/sq. in. (Fig. 6 *a*). As at 600° C., the elastic limit was greater the higher the carbon content, but in this case the variation in initial deformation was much greater, the lowest-carbon steel (0·055%) stretching about 0·2% on loading as against 0·002%, approximately, for the 0·4% carbon steel. Similarly, the rate of creep in the initial stages was correspondingly greater the lower the carbon content. As shown in Fig. 6 (*b*), the maximum time for 0·1% of deformation occurs in a steel with more than 0·4% of carbon. With increasing deformation the maximum shifts to a lower carbon content, until at about 0·4% of deformation the maximum is at approximately 0·25% of carbon.

The testing time was not sufficient to indicate whether the maxima curve shifts to a higher carbon content with increasing deformation, as shown in Fig. 5 (*b*), but it is probable that it does so, although the shift may not be pronounced.

Comparing the time to reach 0·1% of deformation, in Figs. 5 (*b*) and 6 (*b*), it will be noted that the effect of carbon within the limits shown is reversed. This is, of course, due to the fact that in the range of carbon content examined, the curve for 0·1% of deformation lies, in Fig. 5 (*b*), on the right-hand side of the maxima curve, and in Fig. 6 (*b*) on the left-hand side.

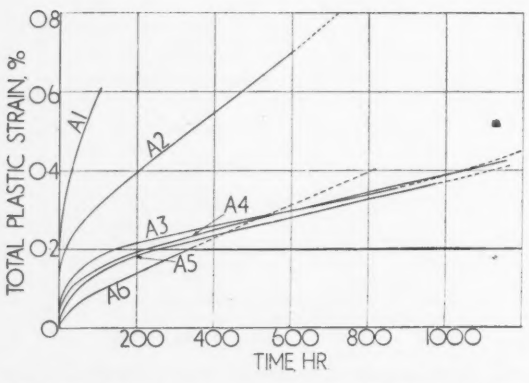


Fig. 6 (*a*)—Curves of creep tests on 0·5% Mo steels with various carbon contents. Stress, 9 tons/sq. in.: Temperature, 550° C. (For key *see* Fig. 5 *a*)

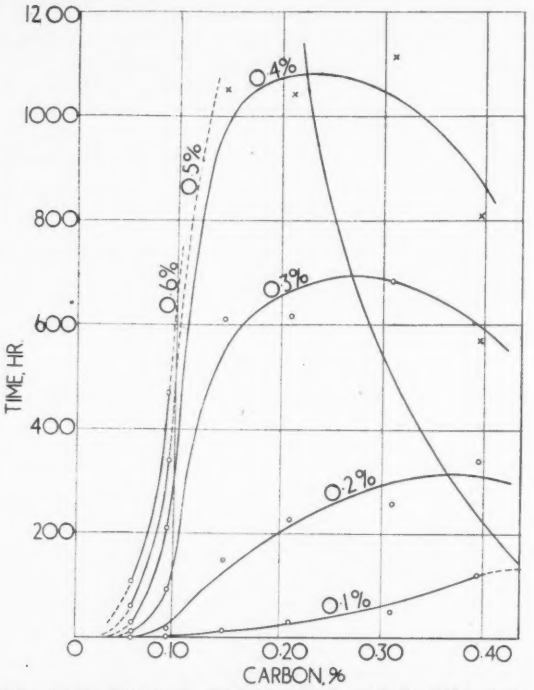


Fig. 6 (*b*)—Effect of carbon content of 0·5% Mo steel A2 on the time to reach the plastic strains indicated. Derived from Fig. 6 (*a*). Stress, 9 tons/sq. in.: Temperature, 550° C.

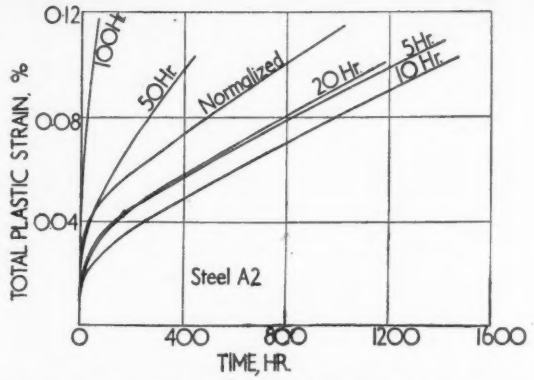


Fig. 7 (*a*)—Curves of creep tests on 0·09%-C/0·5%-Mo steel after tempering for the times indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

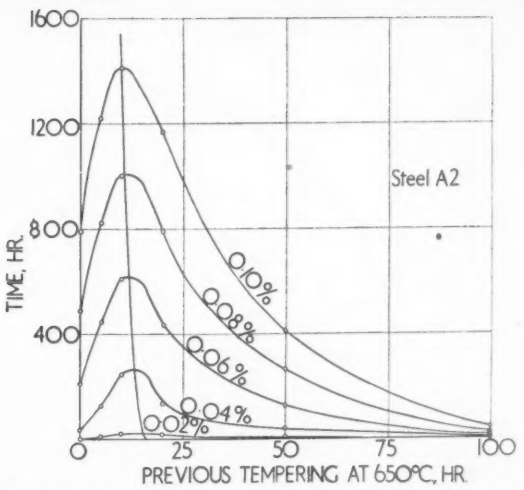


Fig. 7 (*b*)—Effect of time of tempering of 0·09%-C/0·5%-Mo steel A2 on the time to reach the plastic strains indicated. Derived from Fig. 7 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

Creep tests at 550° C. and a stress of 6 tons/sq. in. were carried out on three of the steels. The curves obtained on the normalized steels are shown in Figs. 7 (*a*), 8 (*a*) and 9 (*a*). The effect of carbon content on the time for specific amounts of deformation is shown in Fig. 10. As before, the highest carbon steel is superior in the first few hours of test, but as the deformation and time increase, lower carbon steels become superior. The curve of maximum time reaches a minimum carbon content at about 0·1% of deformation. It must then shift to a higher carbon content because, as shown in Fig. 13 (*a*), in a longer time, the 0·21% carbon steel becomes better than the 0·09% carbon steel.

Under all three sets of test conditions (Figs. 5 *b*, 6 *b* and 10) the maxima curve at small amounts of deformation begins at a certain carbon content, reaches a lower carbon content at a certain deformation, and then shifts or tends to shift to a higher carbon content with further increases in deformation. It is clear, therefore, that the effect of varying the carbon content of normalized 0·5% molybdenum steel follows the same general trend under widely varying test conditions, *but the carbon content which gives the maximum resistance to creep depends on the stress, the temperature, and the amount of permissible deformation.*

In a summary on the effect of carbon content on creep,[9] conflicting results on the effect of up to about 0·5% of carbon are recorded. From Figs. 5 (*b*), 6 (*b*), and 10, it is clear that these conflicting results can be simply explained. Increasing or decreasing creep resistance with increasing carbon content occurs when the test results lie on the left-hand side or the right-hand side, respectively, of the maxima curve for the particular test conditions. At that particular carbon content where the results cross the maxima curve the best creep resistance will be obtained under these conditions.

But the only real criterion of creep resistance is

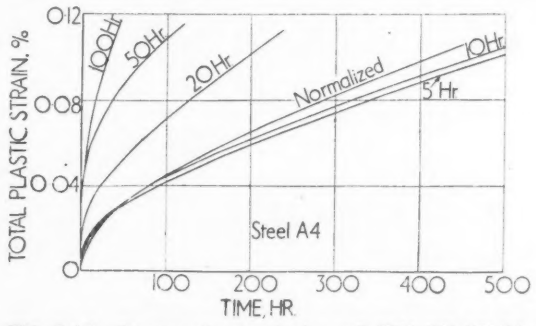


Fig. 8 (*a*)—Curves of creep tests on 0·21%-C/0·5%-Mo steel A4 after tempering for the times indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

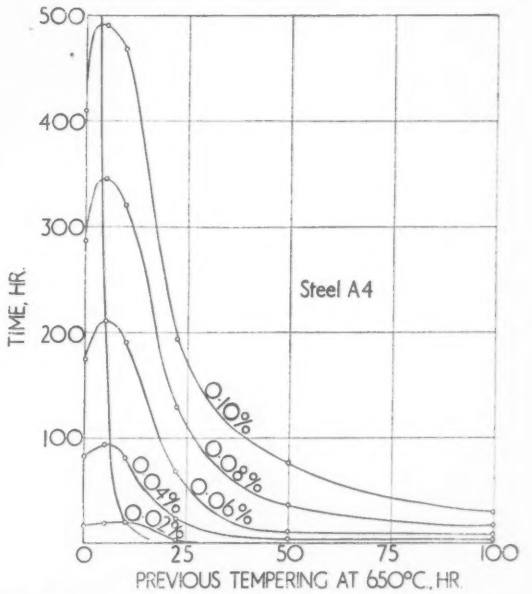


Fig. 8 (*b*)—Effect of tempering time at 650° C. of 0·21%-C/0·5%-Mo steel A4, on the time to reach the plastic strains indicated. Derived from Fig. 8 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

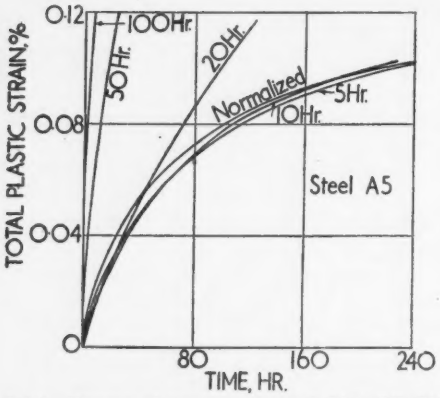


Fig. 9 (*a*)—Curves of creep tests on 0·31%-C/0·5%-Mo steel A5 after tempering at 650° C. for the times indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

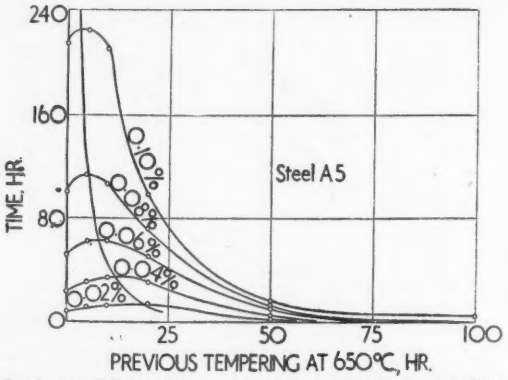


Fig. 9 (*b*)—Effect of tempering time at 650° C. of 0·31%-C/0·5%-Mo steel A5 on the time to reach the plastic strains indicated. Derived from Fig. 9 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

that stress which will give not more than a certain permissible deformation during the working life of the steel. It is necessary to determine whether the test results can indicate which steel will withstand the greatest stress under the working conditions.

It may be said that a steel has two kinds of creep resistance when tested in the normalized condition. An initial resistance, and a latent resistance which develops with time at temperature (the higher the temperature the more quickly does this latent resistance develop and the more quickly does subsequent deterioration set in). The above results on the normalized 0·5% molybdenum steels indicate that a high carbon steel has a greater initial resistance, but a smaller latent resistance than lower carbon steel, but before any conclusions relating to long-time properties can be reached a more accurate estimate of this latent resistance is necessary, and also a further effect of time at temperature must be considered, namely, the deterioration of creep resistance to spheroidization.

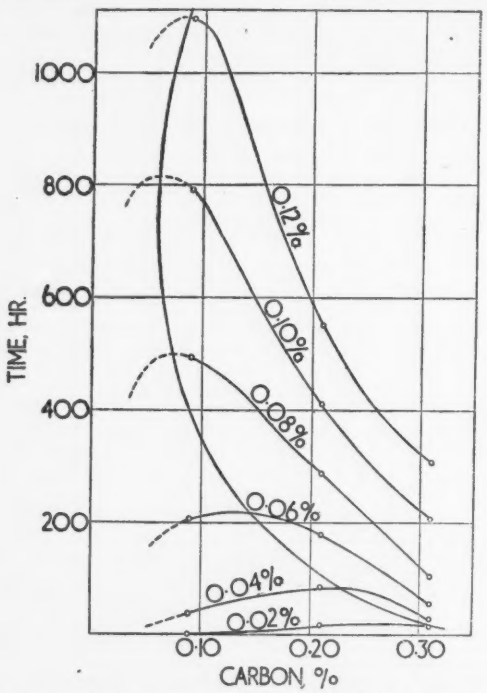


Fig. 10—Effect of carbon content of 0·5% Mo steel on the time to reach the plastic strains indicated. Derived from Fig. 7 (*a*), Fig. 8 (*a*), and Fig. 9 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

EFFECT OF TEMPERING

In the above comparison the tests were necessarily of limited duration and thus do not take full cognizance of the possible deteriorating effect of the time of holding at temperature. Accordingly, three of the steels were tested at a temperature of 550° C. and

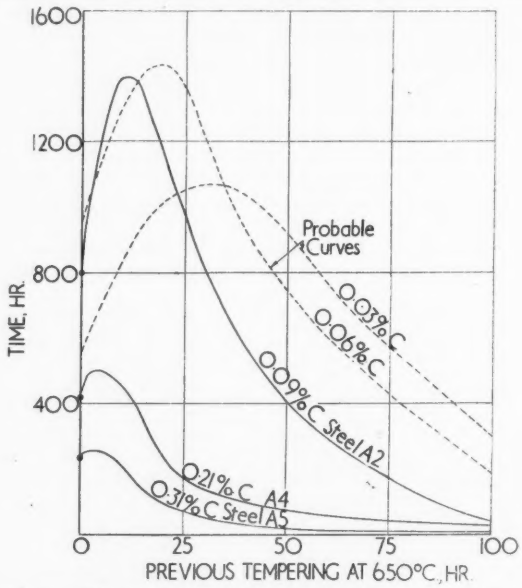


Fig. 11—Effect of time of tempering at 650° C. of various 0·5% Mo steels on the time for 0·1% of plastic strain. Derived from Figs. 7 (*a*), 8 (*a*), and 9 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

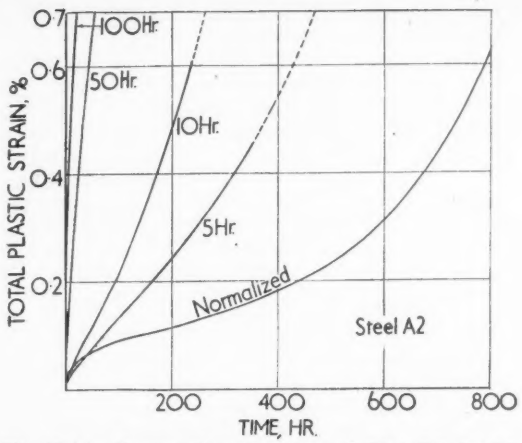


Fig. 12 (*a*)—Curves of creep tests on 0·09%-C/0·5%-Mo steel A2 after tempering at 650° C. for the times indicated. Stress, 4 tons/sq. in.: Temperature, 600° C.

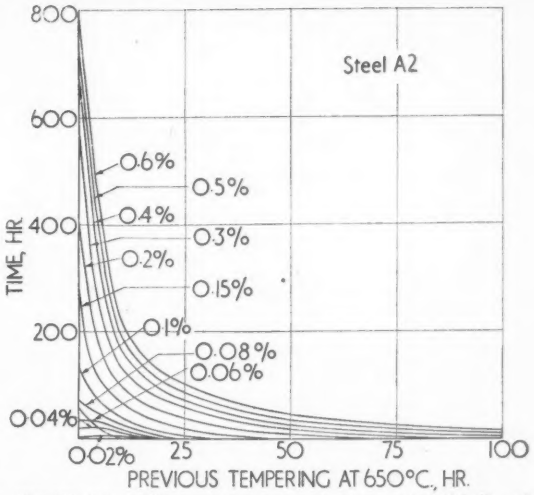


Fig.12 (*b*)—Effect of tempering time at 650° C. of 0·09%-C/0·5%-Mo steel A2 on the time to reach the plastic strains indicated. Derived from Fig. 12 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

a stress of 6 tons/sq. in. after 5, 10, 20, 50, and 100 hr. tempering at 650° C. followed by air cooling.

For the 0·09% carbon steel, Fig. 7 (*a*) shows the creep curves and Fig. 7 (*b*) the curves of time for specific deformation plotted against time of tempering. The maxima in the curves of deformation were obtained after a fairly short tempering time and the maxima curve gradually approaches the zero tempering line with increasing deformation.

Figures 8 (*a* and *b*) show similar curves for the 0·21% carbon steel. In this case, the maxima curve is somewhat nearer the zero tempering line. Similar results are shown in Figs. 9 (*a* and *b*) for the 0·31% carbon steel.

The 0·1%-deformation curves of the 0·09, 0·21, and 0·31% carbon steels after various tempering treatments are reproduced in Fig. 11, which also includes the probable curves for two lower carbon steels. Under these conditions the lower the carbon content the longer the tempering time required to cause maximum stiffening, and also the longer will be the tempering time required to cause serious deterioration.

Similar tests at a temperature of 600° C. and a stress of 4 tons/sq. in. have been carried out on the 0·09% carbon steel again after tempering. The creep curves are shown in Fig. 12 (*a*), and the time for specific deformation plotted against tempering time, in Fig. 12 (*b*). The stiffening effect is indicated only at very small amounts of deformation and short-time tempering. By considering the curve for 0·1% of deformation it can be seen that the steel appears to show no stiffening effect at all due to tempering.

Comparing Figs. 7 (*b*) and 12 (*b*) it is seen that increasing the temperature of the creep tests reduces the tempering time at which maximum stiffening

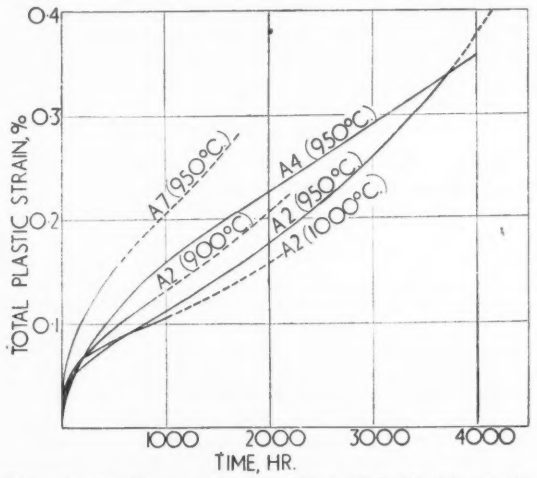


Fig. 13 (*a*)—Curves of creep tests on 0·5% Mo steels A2, A4, and A7, showing the effect of varying the normalizing temperature as indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

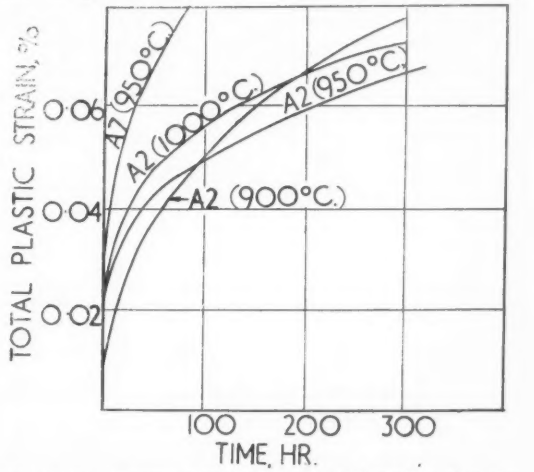


Fig. 13 (*b*)—Curves showing the initial stages of creep tests on 0·5% Mo steels A2 and A7 on the results given in Fig. 13 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

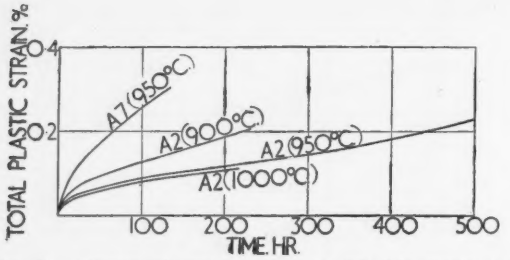


Fig. 14—Curves of creep tests on 0·5% Mo steels A2 and A7 normalized as indicated. Stress, 4 tons/sq. in.: Temperature, 600° C.

occurs. Since in the tests at 600° C. the time was so short that any additive tempering effect due to this time can be neglected, it is clear that Bailey's relationship connecting the rate of spheroidization with change in creep properties is not directly applicable to these steels. This aspect will be dealt with in more detail later.

In Figs. 7 (*b*), 8 (*b*), and 9 (*b*) it is shown that the *maxima curves* gradually approach the zero tempering line with increasing deformation, but the time required may be very long. With increasing stress the time required will be shortened. This can be deduced from the data already discussed and is confirmed by the evidence of rupture tests which are reported later.

EFFECT OF NORMALIZING TEMPERATURE

So far, attention has been confined to the effect of several variables but keeping the normalizing temperature constant. The effect of varying the normalizing temperature has been investigated for the 0·09% carbon steel and is shown by the creep curves at a temperature of 550° C. and a stress of 6 tons/sq. in., in Fig. 13 (*a* and *b*). Three of the curves represent this steel normalized from 900°, 950°, and 1000° C., respectively. A steel of similar composition made in the 18-lb. high-frequency furnace, normalized at 950° C., gave a finer structure and inferior creep properties. The first part of the creep curves is indicated in Fig. 13 (*b*) on a larger scale. With increasing coarseness of structure there is first of all a decrease and then an increase in the amount of deformation in the initial stages. This corresponds with an increase and then a decrease in the elastic limit on loading. With increase in time of test, however, there is a continual cross-over until all the curves align themselves into order, the coarsest structure being the best, *i.e.*, it has the maximum

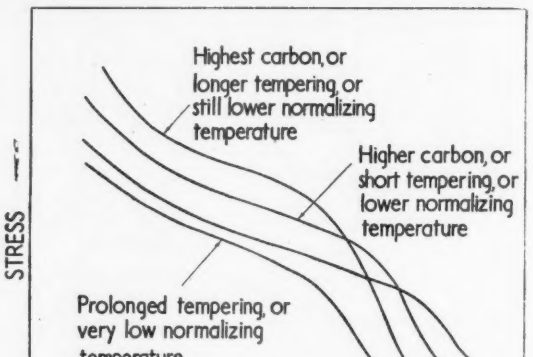

Highest carbon, or longer tempering, or still lower normalizing temperature
Higher carbon, or short tempering, or lower normalizing temperature
Prolonged tempering, or very low normalizing temperature
Lowest carbon, or no tempering, or highest normalizing temperature
STRESS
TEMPERATURE

Fig. 15—Diagrammatic stress/temperature curves of 0·5% Mo steels with different carbon contents, for 0·1% of plastic strain in 100,000 hr.

latent resistance. It is clear that this effect greatly complicates any comparison made between steels of different compositions and heat-treatment, particularly when using a single relatively short-time creep test.

The effect of varying the normalizing temperature was also examined by tests at 600° C. and a stress of 4 tons/sq. in. (Fig. 14). In this case, the cross-over effect occurs in the first few hours of the test, *i.e.*, with increase of testing temperature, the steel normalized from the highest temperature becomes the best more quickly. Increasing the stress will have a similar effect.

A few tests have also been made on the effect of tempering after different normalizing temperatures. These indicate that the lower the normalizing temperature the less the stiffening effect obtained and the shorter the tempering time required to cause maximum stiffening. These results have not been reported, since the effect is more clearly demonstrated by tests on the chromium–molybdenum steels discussed later.

DISCUSSION

Since it has been shown that maximum latent resistance corresponds to minimum deterioration due to spheroidization, it is possible to sum up the

Table II

MANUFACTURING DETAILS AND ANALYSIS OF THE CHROMIUM–MOLYBDENUM STEELS

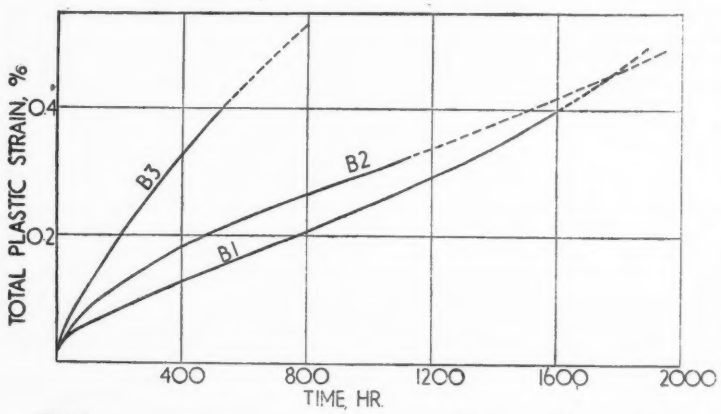
Steel	Process	Method of Deoxidation	Analysis								
			C, %	Si, %	S, %	P, %	Mn, %	Ni, %	Cr, %	Mo, %	Cu, %
B1	Acid open-hearth	SiMn to ladle	0·10	0·14	0·033	0·036	0·43	0·15	0·72	0·53	0·065
B2	,, ,,	SiMn to ladle	0·145	0·15	0·030	0·031	0·53	0·08	0·75	0·54	0·05
B3	600 lb. high-frequency	FeMn and FeSi	0·22	0·23	0·024	0·014	0·63	...	0·48	0·57	...
B4	Acid open-hearth	SiMn to ladle	0·125	0·20	0·031	0·028	0·51	...	0·88	0·515	...
B5	18 lb. high-frequency	FeMn and FeSi	0·101	0·21	0·030	0·019	0·46	0·10	0·82	0·49	0·08
B6	Acid open-hearth	SiMn to ladle	0·12	0·17	0·029	0·030	0·51	...	0·82	0·585	...

effect of carbon on normalized 0·5% molybdenum steels. For 0·1% of deformation at any particular temperature a high carbon steel will be the best at high stresses, but the time will be very short. At a lower stress a lower carbon steel will be the best and the time will be rather longer. With increase of temperature a lower carbon steel becomes the better in a shorter time.

Since an increase in tempering temperature shortens the time for a lower carbon steel to become the better, therefore at a constant time, if a particular steel is best at one temperature a lower carbon steel will be the best at a somewhat higher temperature. Thus, the curves of stress/temperature relationship for 0·1% of deformation in 100,000 hr. for normalized 0·5% molybdenum steels of different carbon contents, will be of the form shown in Fig. 15. If one of these curves is known, then it is possible to indicate whether a steel of different carbon content will be better or worse at a particular temperature. As will be shown later, a similar argument can be applied to the effect of other steel variables.

The effect of tempering on the stress/temperature curve for 0·1% of deformation in 100,000 hr. is more complicated. A short tempering time will have the same effect as increasing the carbon content, *i.e.*, the steel will be better below a certain temperature and worse above that temperature. With more prolonged tempering the steel will be inferior at all temperatures as shown in Fig. 15. In Fig. 2 (*b*) the sharp drop in the stress/temperature relationship for 0·1% of creep in 100,000 hr. occurs between 900° and 950° F. Since this drop is associated with the deterioration due to spheroidization it is clear that in 0·5% molybdenum steels tempering is only beneficial where the service temperature is below 900° F.

The effect of normalizing is also complex. A high normalizing temperature gives better properties above a certain testing temperature, below which a lower normalizing temperature will be better. Very low normalizing temperatures, particularly if the steel is made in an electric furnace, gives inferior properties at all temperatures. This effect also is illustrated in Fig. 15.



Steel	B1	B2	B3
Carbon content, %	0·10	0·145	0·22

Fig. 16—Curves of creep tests on 0·8%-Cr/0·5%-Mo steels with different carbon contents, normalized at 950° C. Stress, 4 tons/sq. in.: Temperature, 600° C.

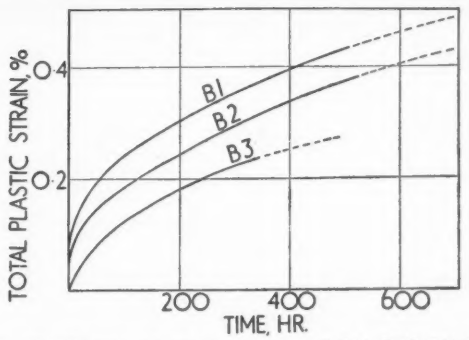


Fig. 17—Curves of creep tests on 0·8%-Cr/0·5%-Mo steels with various carbon contents. Stress 9 tons/sq. in.: Temperature, 550° C. (For key see Fig. 16)

CREEP TESTS ON 0·8%-CHROMIUM/ 0·5%-MOLYBDENUM STEEL

The addition of chromium in amounts up to approximately 1% improves the long-time creep properties of 0·5% molybdenum steel, but greater additions seriously decrease the creep strength. An important effect of added chromium is that the structure is much finer that that of the molybdenum steel after the same normalizing treatment.

A series of tests on similar lines to those described above have been carried out on chromium–molybdenum steel. Although a considerable number of steels made by the open-hearth process were available, these were all of 0·10–0·15% carbon content. A sample of higher carbon content was, however, available from a 600 lb. high-frequency cast of steel. Details of the steels used are listed in Table II.

EFFECT OF CARBON CONTENT

The effect of varying the carbon content was investigated by using three steels of 0·10, 0·145, and 0·22% of carbon, respectively, other elements being essentially the same.

The creep curves at a temperature of 600° C. and a stress of 4 tons/sq. in. are shown in Fig. 16. The test on the 0·145% carbon steel was continued for 1000 hr. only, but the curve of this test has been extrapolated to cross the curve of the 0·1% carbon steel because the higher carbon steel was known to have a better rupture strength at this stress and temperature.

The creep curves at 550° C. and a stress of 9 tons/sq. in. are shown in Fig. 17, and those at the same temperature but at a stress of 6 tons/sq. in. are shown in Fig. 18.

Although only three tests were

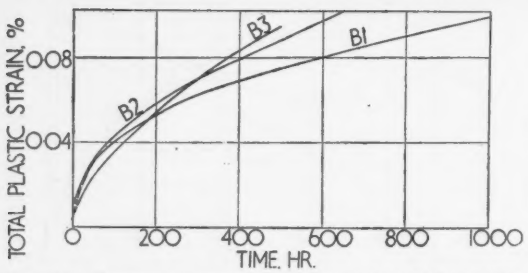


Fig. 18—Curves of creep tests on 0·8%-Cr/0·5%-Mo steels with various carbon contents. Stress, 6 tons/sq. in.: Temperature, 550° C. (For key *see* Fig. 16)

carried out under each set of conditions the results indicate that the effect of varying the carbon content of chromium–molybdenum steel is similar to that on 0·5% molybdenum steel, and the same conclusions are applicable. It will be noted that the chromium–molybdenum steels are superior to the corresponding 0·5% molybdenum steels. This is particularly true when the carbon content is about 0·10%.

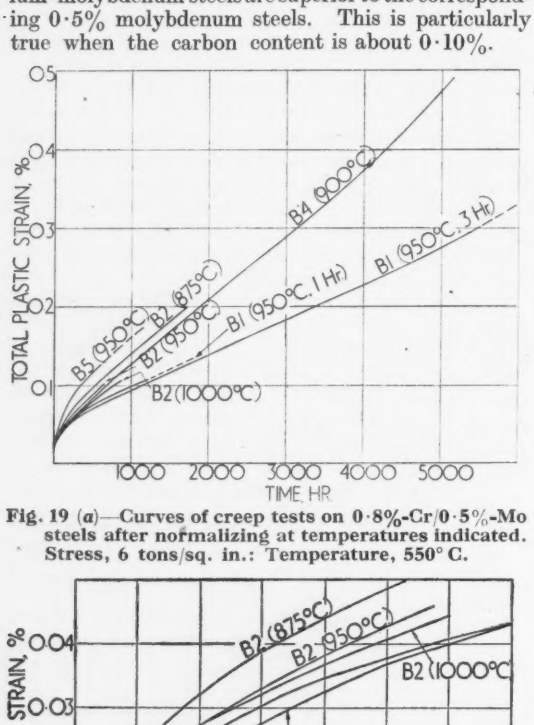


Fig. 19 (*a*)—Curves of creep tests on 0·8%-Cr/0·5%-Mo steels after normalizing at temperatures indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

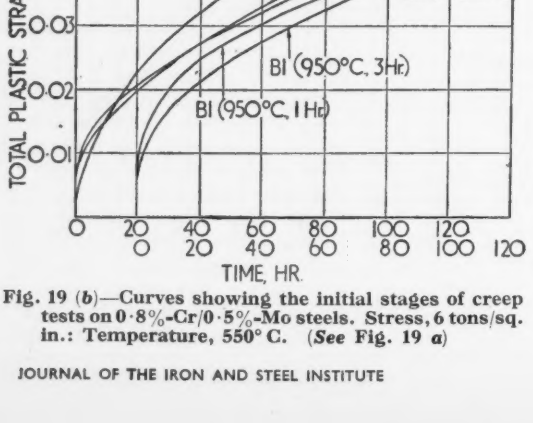


Fig. 19 (*b*)—Curves showing the initial stages of creep tests on 0·8%-Cr/0·5%-Mo steels. Stress, 6 tons/sq. in.: Temperature, 550° C. (*See* Fig. 19 *a*)

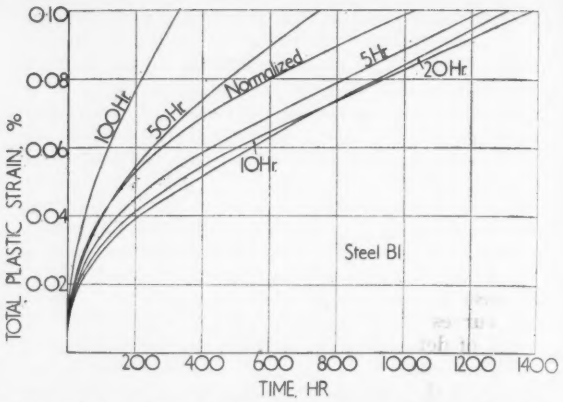


Fig. 20 (*a*)—Curves of creep tests on 0·1%-C/Cr/Mo steel B1 (normalized 1 hr. at 950° C.) after tempering at 650° C. for the times indicated on the curves. Stress, 6 tons/sq. in.: Temperature, 550° C.

EFFECT OF NORMALIZING TEMPERATURES

The effect of varying the normalizing temperature has been studied by creep tests at a temperature of 550° C. and a stress of 6 tons/sq. in., as shown in Fig. 19 (*a* and *b*). With increasing normalizing temperature there is first of all a decrease and then an increase in the amount of deformation in the initial stages, the effect being similar to that found with the 0·5% molybdenum steels. At larger deformations a considerable improvement in the creep resistance is obtained by increasing the normalizing temperature. Within the limits of the test the 0·1% carbon steel is superior to the 0·145% carbon steel but, as shown later, the higher carbon steel has a superior rupture strength at this temperature and stress. The reason for this is that the lower carbon steel enters the third phase of creep at a smaller deformation, and the two curves cross at a later stage of creep. This effect also occurred at 600° C. and 4 tons/sq. in. as already discussed (Fig. 16).

The test on the high-frequency-furnace cast of 0·10% carbon steel showed it to have a poorer

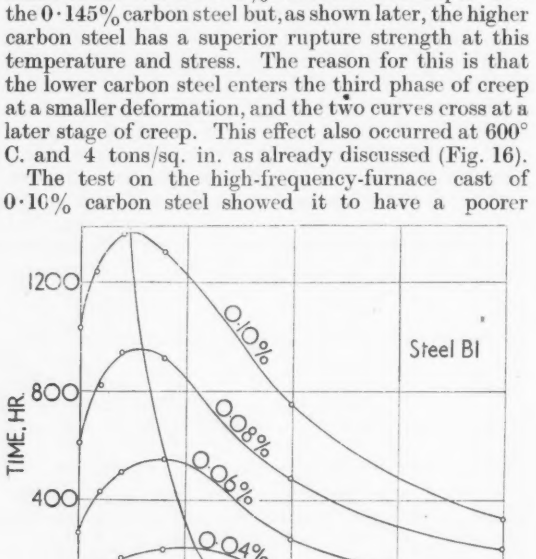


Fig. 20 (*b*)—Effect of tempering time at 650° C. of 0·1%-C/Cr/Mo steel B1 (normalized 1 hr. at 950° C.) on the time to reach the plastic strains indicated. Derived from Fig. 20 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

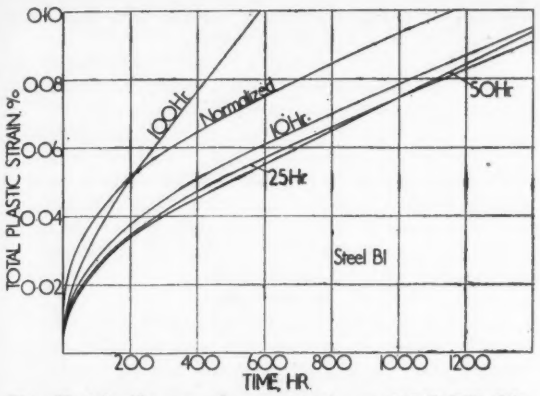


Fig. 21 (*a*)—Curves of creep tests on 0·1%-C/Cr/Mo steel B1 (normalized 3 hr. at 950° C.) after tempering at 650° C. for the times indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

creep resistance than the corresponding open-hearth steel. Steel made in a 5-ton electric arc furnace is better but is still inferior to open-hearth steel. A similar effect was observed with 0·5% molybdenum steel (Fig. 13 *a* and *b*). Although normalized at the same temperature as the open-hearth steel, the electric-furnace steels had a finer structure.

EFFECT OF TEMPERING

The effect of tempering at 650° C. after two different normalizing treatments was investigated using 0·1% carbon steel. The results after pre-treatment at 950° C. for 1 hr. and air cooling are shown in Figs. 20 (*a* and *b*). Compared with the corresponding molybdenum steel (Fig. 7 *b*) the stiffening effect is more pronounced and the deterioration is not so marked even after 100 hr. tempering.

The same steel was tempered and re-tested after normalizing at 950° C. for 3 hr. and air cooling (Fig. 21 *a* and *b*). In this case a very pronounced stiffening effect occurred and the deterioration was still less. Even after 100 hr. tempering, the time to reach 0·1% of creep was 580 hr. as against 45 hr. for the corresponding molybdenum steel (Fig. 7 *b*). Some improvement might have been affected in this latter steel, but this could only be achieved by excessive coarsening of the structure and, as shown later, this may lead to premature failure in rupture tests.

The chromium–molybdenum steel normalized at 950° C. (3 hr.) was also tested at 4 tons/sq. in. and 600° C. after various tempering treatments, as shown

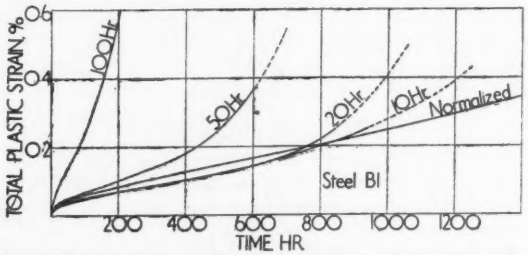


Fig. 22 (*a*)—Curves of creep tests on 0·1%-C/Cr/Mo steel B1 (normalized 3 hr. at 950° C.) after tempering at 650° C. for the times indicated. Stress, 4 tons/sq. in.: Temperature, 600° C.

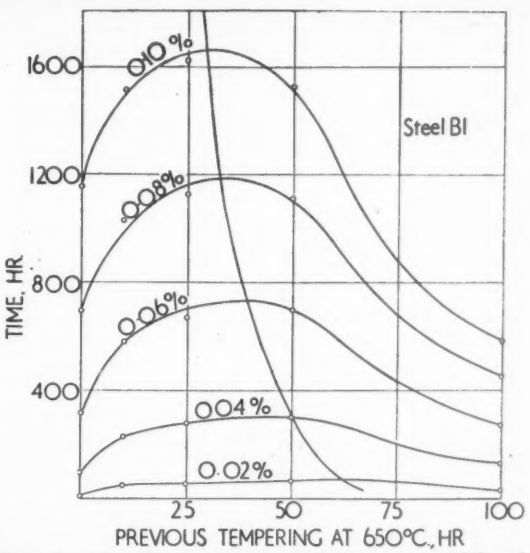


Fig. 21 (*b*)—Effect of tempering time at 650° C. of 0·1%-C/Cr/Mo steel B1 (normalized 3 hr. at 950° C.) on the time to reach the plastic strains indicated. Derived from Fig. 21 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

in Fig. 22 (*a* and *b*). The stiffening effect is still quite pronounced even up to 0·2% of deformation, but at about 0·3% of deformation or more the test on the untempered steel is the best because the tempered tests enter the third stage of creep more quickly. The molybdenum steel (Fig. 12 *b*) under corresponding conditions showed practically no stiffening at all.

Further data on the effect of tempering is shown in Fig. 23, the creep curves of the normalized steels used being shown in Fig. 19 (*a* and *b*). Increasing the normalizing temperature increased the degree of stiffening, increased the tempering time at which

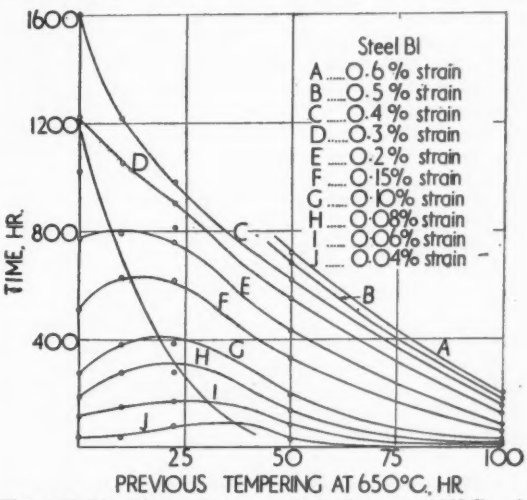


Fig. 22 (*b*)—Effect of tempering time at 650° C. of 0·1%-C/Cr/Mo steel B1 (normalized 3 hr. at 950° C.) on the time to reach the plastic strains indicated. Stress, 4 tons/sq. in.: Temperature, 600° C.

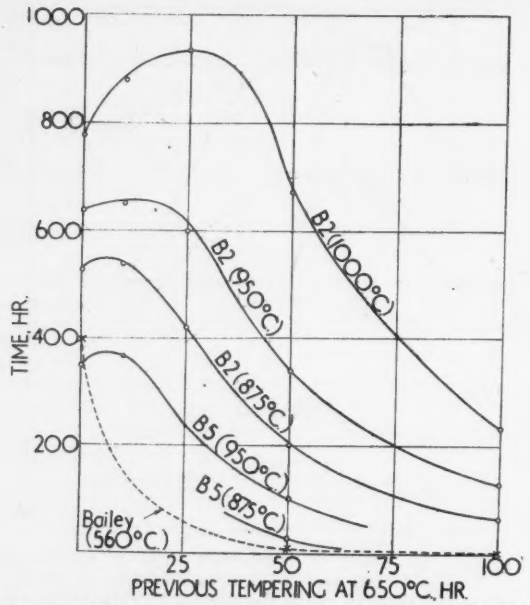


Fig. 23—Effect of tempering time at 650° C. of Cr–Mo steels B2 and B5 (normalized as indicated) on the time to reach 0·1% of plastic strain

the maximum stiffening was obtained, and decreased the amount of deterioration obtained at any stage. The 0·145% carbon steel was not so good as the 0·1% carbon steel (Figs. 20 *b* and 21 *b*).

The high-frequency cast of 0·10% carbon steel was inferior to the open-hearth steel both in the normalized condition and after tempering (Fig. 23). At 50 hr. tempering the high-frequency steel was tested in two conditions of normalizing, namely, 950° and 875° C., the results in the latter condition being the worse. Bailey[10] has quoted results for 0·5%-molybdenum/1%-chromium steel but the full composition and method of heat-treatment were unfortunately not stated. The results obtained have been included in Fig. 23, and indicated very poor properties for chromium–molybdenum steel. These tests were carried out at a temperature of 560° C., but the difference of 10° is not sufficient to account for the very poor properties. The results would indicate that Bailey's steel was made in the electric furnace or a high-frequency furnace and had been normalized at a low temperature.

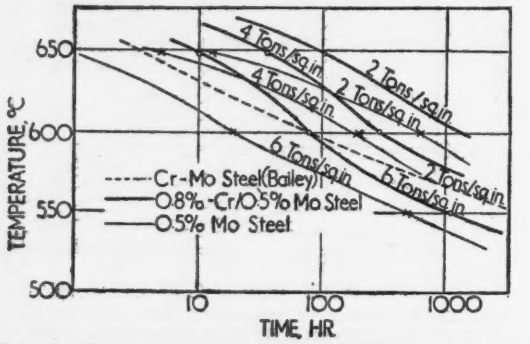


Fig. 24—Temperature/time curves of Mo and Cr–Mo steels for 0·1% plastic strain at stresses indicated

Further results quoted by Bailey are shown in Fig. 24 and are compared with a 0·10% carbon open-hearth cast of chromium–molybdenum steel. The results for the 0·09%-carbon/0·5%-molybdenum open-hearth steel are also included, and the pronounced superiority of the open-hearth chromium–molybdenum steel is evident. It is clear that comparisons between steels of different composition must be made with great care, otherwise a class of steel may be condemned without justification.

DISCUSSION

It has been shown that the chromium–molybdenum steel is influenced in the same way as 0·5% molybdenum steel by variation of the testing conditions, the heat-treatment, and the carbon content. Thus, the points made in the discussion of the effect of these variables on molybdenum steel also apply to chromium–molybdenum steel.

In open-hearth steels the addition of chromium refines the structure of 0·5% molybdenum steel, an effect which tends to decrease the creep strength at high temperature. If the steels had been compared using different treatments to give similar structures, the chromium–molybdenum steel would have shown up to still greater advantage.

EFFECT OF ANNEALING ON CREEP PROPERTIES

Furnace cooling from the normalizing temperature is sometimes employed, particularly for castings. In most cases this is very detrimental to long-time creep properties, although for some steels it may appear to be beneficial in a short-time test because the slow cooling through the tempering range may cause stiffening which would make the steel appear good in this test. However, a test, after tempering at say 650° C., would show that the steel would deteriorate rapidly in service. The effect of furnace cooling of molybdenum and chromium–molybdenum steel is shown in Fig. 25.

In the foregoing work an attempt has been made to indicate the effect of carbon, normalizing temperature, and tempering time on the creep resistance both in relation to each other and also with varying stress, temperature, and duration of test. It now remains to discuss the effect of the deoxidizing elements aluminium, silicon, and manganese. Details of the steels used in this part of the investigation are given in Table III.

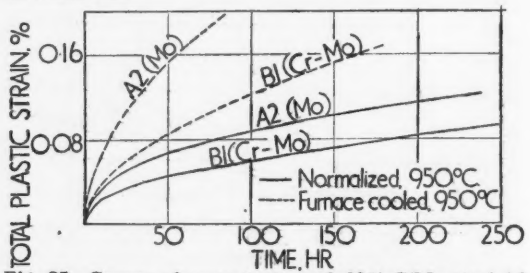


Fig. 25—Curves of creep tests on 0·09%-C/Mo steel A2 and 0·1%-C/Cr/Mo steel B1 (normalized or annealed at 950° C.). Stress, 4 tons/sq. in.: Temperature, 600° C.

Table III

EFFECT OF ADDITIONS OF ALUMINIUM, SILICON, AND MANGANESE TO THE STEELS

Steel	Process	Method of Deoxidation	Analysis								
			C, %	Si, %	S, %	P, %	Mn, %	Ni, %	Cr, %	Mo, %	Cu %
D1	18 lb. high-frequency furnace	FeMn, FeSi, and 2 lb. Al/ton of steel	0·115	0·16	0·021	0·009	0·40	0·04	Tr.	<0·01	Tr.
D2	,, ,, ,,	,, ,, ,,	0·115	0·18	0·021	0·009	0·59	0·04	Tr.	0·01	Tr.
D3	,, ,, ,,	,, ,, ,,	0·11	0·15	0·021	0·009	0·79	0·04	Tr.	0·01	Tr.
D4	,, ,, ,,	,, ,, ,,	0·12	0·16	0·021	0·009	1·52	0·04	Tr.	0·01	Tr.
D5	,, ,, ,,	FeMn and FeSi	0·11	0·14	0·021	0·009	0·28	0·04	Tr.	0·01	Tr.
D6	,, ,, ,,	,, ,,	0·115	0·17	0·021	0·009	0·39	0·04	Tr.	0·01	Tr.
D7	,, ,, ,,	,, ,,	0·125	0·15	0·021	0·009	0·49	0·04	Tr.	0·01	Tr.
D8	,, ,, ,,	,, ,,	0·12	0·16	0·021	0·009	0·88	0·04	Tr.	0·01	Tr.
E 1	,, ,, ,,	FeMn, FeSi, and 3 lb. Al/ton of steel	0·14	0·18	0·021	0·009	0·57	0·04	Tr.	<0·01	Tr.
E 2	,, ,, ,,	,, ,, ,,	0·15	0·16	0·021	0·009	0·72	0·04	Tr.	0·05	Tr.
E 3	,, ,, ,,	,, ,, ,,	0·16	0·17	0·021	0·009	0·72	0·04	Tr.	0·11	Tr.
E 4	,, ,, ,,	,, ,, ,,	0·145	0·16	0·021	0·009	0·67	0·04	Tr.	0·15	Tr.
E 5	,, ,, ,,	,, ,, ,,	0·13	0·16	0·021	0·009	0·46	0·04	Tr.	0·20	Tr.
E 6	,, ,, ,,	,, ,, ,,	0·125	0·18	0·021	0·009	0·71	0·04	Tr.	0·32	Tr.
E 7	,, ,, ,,	,, ,, ,,	0·13	0·17	0·021	0·009	0·66	0·04	Tr.	0·38	Tr.
E 8	,, ,, ,,	,, ,, ,,	0·135	0·18	0·021	0·009	0·73	0·04	Tr.	0·49	Tr.
E 9	,, ,, ,,	FeMn and FeSi	0·14	0·16	0·021	0·009	0·68	0·04	Tr.	0·01	Tr.
E10	,, ,, ,,	,, ,, ,,	0·12	0·15	0·021	0·009	0·65	0·04	Tr.	0·20	Tr.
E11	,, ,, ,,	,, ,, ,,	0·145	0·15	0·021	0·009	0·71	0·04	Tr.	0·30	Tr.
E12	,, ,, ,,	,, ,, ,,	0·14	0·16	0·021	0·009	0·58	0·04	Tr.	0·37	Tr.
E13	,, ,, ,,	,, ,, ,,	0·14	0·16	0·021	0·009	0·72	0·04	Tr.	0·48	Tr.
F1	,, ,, ,,	FeMn and FeSi	0·115	0·21	0·027	0·013	0·68	0·10	1·03	0·52	0·08
F2	,, ,, ,,	FeMn, FeSi, and 2 lb. Al/ton of steel	0·105	0·23	0·027	0·013	0·52	0·10	0·96	0·51	0·08
H1	600 lb. high-frequency furnace	FeMn and FeSi	0·07	0·06	0·023	0·014	0·33	...	0·72	0·62	...
H2	,, ,, ,,	,, ,,	0·07	0·23	0·023	0·014	0·35	...	0·75	0·62	...
H3	,, ,, ,,	,, ,,	0·065	0·38	0·022	0·015	0·37	...	0·77	0·54	...
H4	,, ,, ,,	,, ,,	0·085	0·75	0·023	0·015	0·42	...	0·81	0·58	...
J1	18 lb. high-frequency furnace	FeMn	0·11	0·01	0·027	0·013	0·62	0·08	0·06	0·03	0·085
J2	,, ,, ,,	FeMn and FeSi	0·105	0·16	0·027	0·013	0·68	0·08	0·06	0·03	0·085
J3	,, ,, ,,	,, ,,	0·095	0·30	0·027	0·013	0·72	0·08	0·06	0·03	0·085
J4	,, ,, ,,	,, ,,	0·095	0·63	0·027	0·013	0·75	0·08	0·06	0·03	0·085
J5	,, ,, ,,	,, ,,	0·125	0·97	0·027	0·013	0·72	0·08	0·06	0·03	0·085

EFFECT OF ALUMINIUM ON CREEP PROPERTIES

Considering the carbon steels shown in Fig. 2 (*a*), it is seen that one of these has very inferior long-time creep properties. Such a steel has been called abnormal and the effect is now known to be due to deoxidation with excess aluminium so that the steel is fine grained in the McQuaid-Ehn test.

For tube manufacture it is desirable that the steel be killed, but as has been shown by the author[11] the extent of such deoxidation need not be so drastic as to cause poor creep properties. It was also shown that the abnormality in carbon steels as measured by a short-time test may be reduced considerably by increasing the manganese content or by the addition of a comparatively small amount of molybdenum (of the order of 0·1%). Increasing the carbon content also markedly reduces the abnormality. The criterion of creep resistance used was the creep rate after 5 days at a temperature of 450° C. and a stress of 8 tons/sq. in.

Such a short-time creep test is of value for certain steels only. In fact, in this test, an aluminium-killed 0·5% molybdenum steel has a better creep resistance than the same steel without the aluminium addition. An illustration of these effects is shown in Fig. 26. The steels used for this experiment were all made in an 18 lb. high-frequency furnace and normalized to give a uniformly fine structure.

The effect of aluminium in carbon steels is to

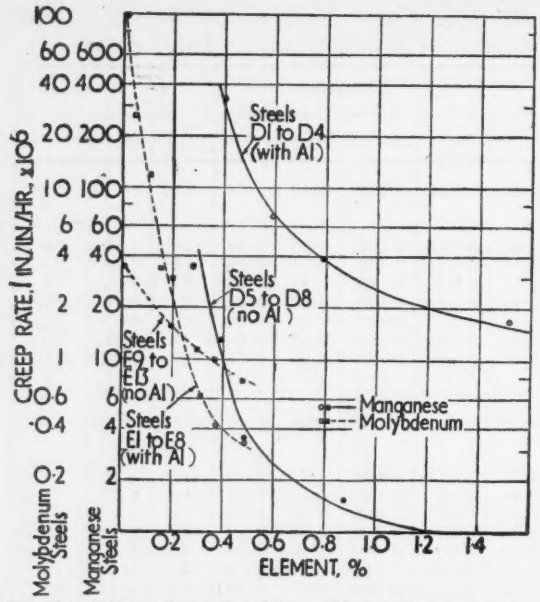


Fig. 26—Effect of varying Mn or Mo on the creep rate of carbon steels deoxidized with and without Al. Stress, 8 tons/sq. in.: Temperature, 450° C.

reduce stiffening and to make it occur at a lower temperature range. Also, deterioration sets in at a lower temperature compared with other steels. The creep strength is consequently poor. Although aluminium may have a drastic effect on carbon steels, its effect on alloy steels is not necessarily so pronounced, particularly on the basis of short-time tests, and indeed it may not be detrimental at all in some alloy steels. Since the addition of aluminium tends to maintain a fine structure, even if it had no other effect, its use is not to be recommended, as a fine structure tends to give poor strength at high temperature.

As shown in Fig. 26, aluminium in molybdenum steels improves the steel in a short-time creep test at 8 tons/sq. in. and 450° C. Even in a short-time test at the higher temperature of 550° C. and a stress of 6 tons/sq. in., aluminium may cause a slight improvement in a steel when tested in the

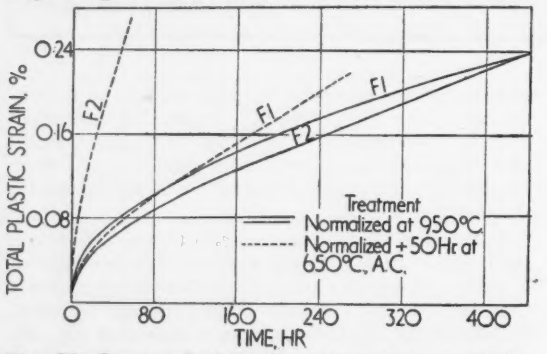
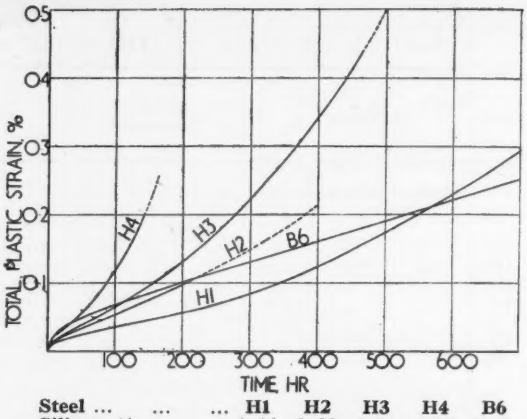


Fig. 27—Curves of short-time creep tests on SiMn-killed (F1) and Al-killed (F2) 0·1%-C/1·0%-Cr/0·5%-Mo steels (normalized at 950° C., and normalized at 950° C. plus 50 hr. at 650° C., A.C.). Stress, 6 tons/sq. in.: Temperature, 550° C.



Steel	H1	H2	H3	H4	B6
Silicon, %	0·06	0·23	0·38	0·75	0·17

Fig. 28 (*a*)—Curves of creep tests on Cr/Mo steels (normalized at 950° C.) with various Si contents. Stress, 4 tons/sq. in.: Temperature, 600° C.

normalized condition. Such a result is shown in Fig. 27 for a 0·5%-molybdenum/1%-chromium steel made in a small high-frequency furnace. In the early stages of the test the aluminium-killed steel in the normalized condition deforms to a lesser extent, but at a later stage it is inferior to the steel without aluminium. This effect is similar to that of a steel tested after different normalizing treatments. At a lower temperature and/or stress, the aluminium-killed steel would be superior over a very much longer period of time. After a tempering treatment of 50 hr. at 650° C., the aluminium steel is from the start very much inferior to the same steel without aluminium. Both the above tests show that in the normalized condition the aluminium-treated steel

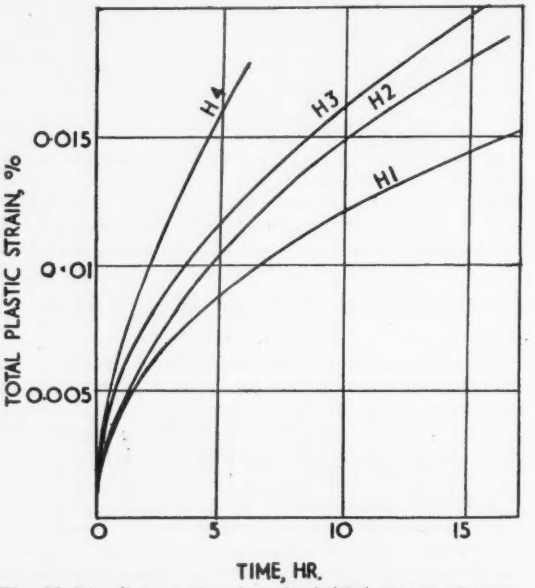


Fig. 28 (b)—Curves showing the initial stages of creep tests given in Fig. 28 (*a*) on Cr/Mo steels (normalized at 950° C.) with various Si contents. Stress, 4 tons/sq. in.: Temperature, 600° C.

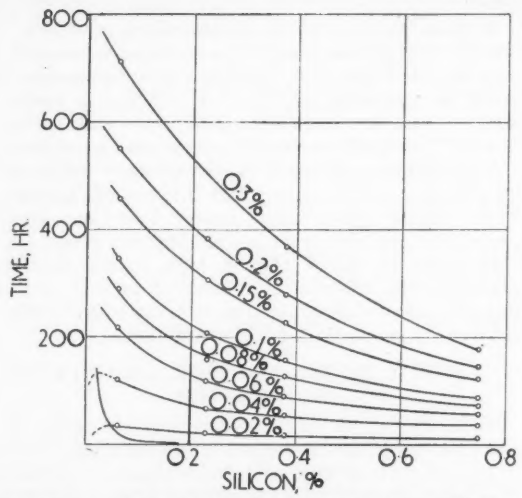


Fig. 28 (*c*)—Effect of Si in Cr–Mo steels (normalized at 950° C.) with various Si contents, on the time to reach the plastic strains indicated. Stress, 4 tons/sq. in.: Temperature, 600° C.

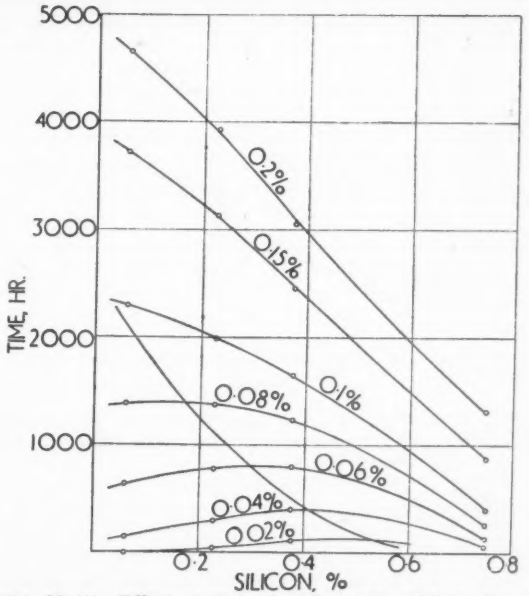


Fig. 29 (*b*)—Effect of Si in Cr–Mo steels with various Si contents on the time to reach the plastic strains indicated. Derived from Fig. 29 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

has a better initial resistance but an inferior latent resistance. Again, a molybdenum–vanadium steel in a short-time test at 550° C. may actually be improved slightly by the addition of aluminium even after a previous tempering treatment.

Since aluminium accelerates the improvement and subsequent deterioration of some, if not all, steels it must be beneficial below a certain temperature depending on the steel. For low carbon steels this temperature is of the order of 600° F. or less[11], at which temperature, however, creep is of little or no importance. Similarly, for molybdenum and chromium–molybdenum steels the temperatures at which the aluminium-killed steel becomes the better is much below that at which the creep is important.

This effect is not confined to aluminium. The rate of improvement and subsequent deterioration can be influenced by varying most of the elements added to steel. For example, silicon has considerable influence depending on the stress, temperature, and time at which the comparison is made.

EFFECT OF SILICON ON CREEP PROPERTIES

The following tests were made on four low carbon chromium–molybdenum steels, the only variable being silicon. This was achieved by making a 600 lb. high-frequency cast and adding ferro-silicon to the furnace between the pouring of each ingot. The steels, in the form of 1-in. dia. bar, were normalized from 950° C., the resulting structure being comparatively fine. For comparison, tests on a normal open-hearth production cast of chromium–molybdenum steel are also shown.

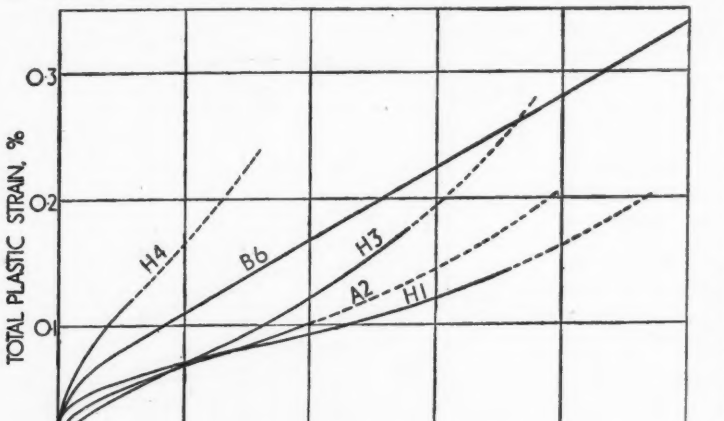


Fig. 29 (a)—Curves of creep tests on Cr–Mo steels with various Si contents. Stress, 6 tons/sq. in.: Temperature 550° C.

In Fig. 28 (*a* and *b*) showing creep tests at a temperature of 600° C. and a stress of 4 tons/sq. in., the four steels place themselves in order of silicon content, the lowest silicon-containing steel showing the minimum deformation at all stages except in the first hour or so of the tests. All four steels enter the third stage of creep at a deformation of less than 0·05%. The open-hearth chromium–molybdenum steel deforms more quickly in the initial stages, but does not enter the third stage until reaching approximately 0·2% of deformation, and the increase in creep rate is more gradual. In Fig. 28 (*c*) silicon content has been plotted

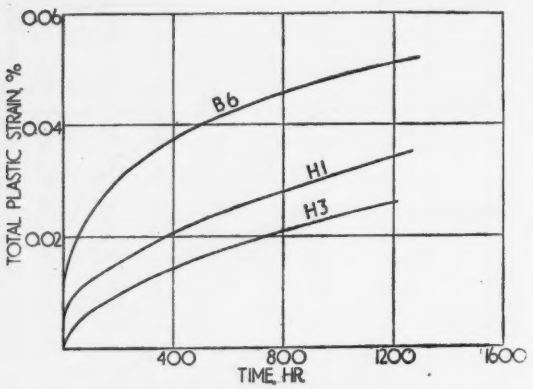


Fig. 30—Curves of creep tests on Cr–Mo steels with various Si contents. Stress, 5·15 tons/sq. in.: Temperature, 538° C.

against the time to reach specified amounts of deformation. For very small deformations a maximum occurs in the curves at a definite silicon content. The maxima curve very rapidly approaches zero silicon content.

In Fig. 29 (*a*), creep tests are shown at a temperature of 550° C. and a stress of 6 tons/sq. in. The trend of the curves is the same, but the cross-over effect which was almost unnoticed in the tests at 4 tons/sq. in. and 600° C. is more pronounced as shown by the maxima curve in Fig. 29 (*b*). In this case a much longer time is required before the lower silicon steels become the best from the point of view of deformation.

In Fig. 30 tests are shown at a temperature of 538° C. and a stress of 5·15 tons/sq. in. In this case, the higher silicon steel is superior both in creep rate and in the amount of deformation up to the limit of the tests (1200 hr.). The open-hearth steel shows considerably greater deformation in the initial stages.

Figure 31 is a summary of further tests at a temperature of 550° C. and a stress of 6 tons/sq. in. after tempering treatments of up to 100 hr. at 650° C., the time to reach 0·1% of deformation being plotted against time of tempering. The most noticeable feature is the apparent lack of any stiffening effect of tempering of the four high-frequency furnace steels compared with the open-hearth steel, and also the rapid drop in the creep properties of the high-frequency furnace steels on tempering. Due to the low carbon and low manganese contents and the comparatively fine structure of these steels the stiffening effect was confined to very small amounts of deformation and thus is not indicated in Fig. 31. The steels compare very unfavourably with the open-hearth steel. At a lower temperature and/or lower stress, a stiffening effect would have been noticed after a few hours tempering. Lowering the normalizing temperature to 900° C. lowers the creep properties considerably, as shown by the tests on the low silicon steel. On the other hand, coarsening of the structure would have lowered the time to reach 0·1% of deformation as compared with the normalized at 950° C. test, but the steels would show a greater degree of stiffening on tempering when tested at a temperature of 550° C. and a stress of 6 tons/sq. in.

A more detailed survey of the effect of silicon is nearing completion and it can be said that the data depicted above for chromium–molybdenum steel apply in a similar way to ordinary 0·5% molybdenum steel. The same effect also occurs with carbon steels. This is illustrated by creep tests carried out at a temperature of 450°C., and a stress of 8 tons/sq. in. (Fig. 32 *a* and *b*), from which it is evident that the lower silicon steels give better long-time creep properties and better long-time rupture properties.

It must be noted that a high silicon content augments the effect of aluminium if present, and a fine-grained steel is obtained more easily with resulting poor creep properties at high temperatures.

EFFECT OF MANGANESE ON CREEP PROPERTIES

From Fig. 26 it will be noted that increasing the manganese content increases the creep resistance of 0·1% carbon steel both with and without the addition of aluminium, and it is interesting to note the very pronounced effect of manganese in the lower range, *i.e.*, manganese acts like carbon and improves the creep resistance under high stresses. At a lower stress the effect of manganese would not have been so pronounced, but the results do indicate that in carbon steel for high-temperature service the manganese content should not be too low.

Further work in progress indicates that variation in manganese content has an appreciable effect also on molybdenum and chromium–molybdenum steels. This indicates that manganese added to 0·5% molybdenum steel has an effect similar to but less pronounced than that of chromium.

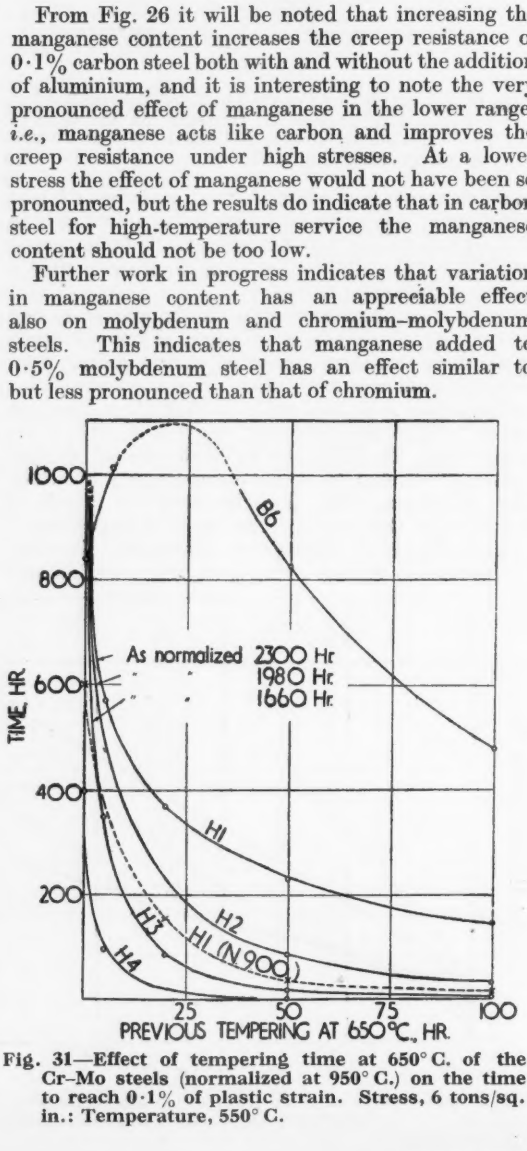


Fig. 31—Effect of tempering time at 650° C. of the Cr–Mo steels (normalized at 950° C.) on the time to reach 0·1% of plastic strain. Stress, 6 tons/sq. in.: Temperature, 550° C.

Table IV

MANUFACTURING DETAILS AND ANALYSIS OF THE MOLYBDENUM–VANADIUM STEELS

Steel	Process	Method of Deoxidation	Analysis								
			C, %	Si, %	S, %	P, %	Mn, %	Ni, %	Cr, %	Mo, %	V, %	Cu, %
K 1	18 lb. high-frequency furnace	FeMn and FeSi	0·10	0·06	0·030	0·018	0·41	0·10	0·07	0·27	0·0	0·085
K 2	600 lb. ,, ,,	,, ,,	0·08	0·04	0·025	0·012	0·39	0·08	0·02	0·29	0·11	0·04
K 3	,, ,, ,,	,, ,,	0·085	0·03	0·029	0·014	0·40	0·08	0·02	0·29	0·21	0·04
K 4	,, ,, ,,	,, ,,	0·08	0·05	0·031	0·016	0·42	0·08	0·02	0·30	0·32	0·04
K 5	18 lb. ,, ,,	FeMn and FeSi	0·11	0·14	0·030	0·019	0·45	0·10	0·07	0·58	0·0	0·085
K 6	600 lb. ,, ,,	,, ,,	0·105	0·12	0·034	0·021	0·31	0·06	0·01	0·60	0·10	0·05
K 7	,, ,, ,,	,, ,,	0·115	0·11	0·033	0·020	0·49	0·06	0·01	0·595	0·215	0·05
K 8	,, ,, ,,	,, ,,	0·12	0·13	0·030	0·019	0·31	0·06	0·01	0·58	0·31	0·05
K 9	7½ ton electric furnace	FeMn, FeSi, and 1 lb. Al to furnace before tap	0·19	0·24	0·020	0·031	0·51	..	..	0·53	0·31	..
K10	,, ,, ,,	FeMn, FeSi, and 1 lb. Al to furnace before tap, and 2 lb. Al to ladle	0·13	0·30	0·019	0·015	0·48	0·09	0·05	0·62	0·31	0·04
K11	600 lb. high-frequency furnace	FeMn and FeSi	0·08	0·03	0·030	0·009	0·40	0·10	0·12	0·46	0·26	0·07
K12	Basic open-hearth	FeSi, SiMn, and 1 lb. Al/ton to ladle	0·12	0·16	0·028	0·011	0·57	0·21	0·14	0·54	0·265	0·05

CREEP TESTS ON MOLYBDENUM–VANADIUM STEEL

The addition of 0·1% or more of vanadium to 0·5% molybdenum steel gives the steel a very fine structure even at comparatively high normalizing temperatures. A McQuaid-Ehn test also shows the steel to be fine-grained (6–8 grain-size), *i.e.*, the steel is thoroughly deoxidized by the vanadium addition. The addition of aluminium, therefore, has not the same detrimental effect as it has on carbon steel, 0·5% molybdenum steel, or chromium–molybdenum steel. Its only effect is to maintain a very fine structure at high normalizing temperatures. For this reason the use of aluminium is not to be recommended.

For the same reason the method of manufacture has not such a strong influence on molybdenum–vanadium steel as it has on molybdenum or chromium–molybdenum steel. Steels made by the open-hearth process, however, are still superior to those made in the electric furnace or high-frequency furnace.

The following data show the pronounced superiority of molybdenum–vanadium steel over the steels considered above, and also show that the characteristic behaviour of molybdenum–vanadium steel is merely an enhancement of the effects which have already been demonstrated for the other steels. The particulars of the steels used are shown in Table IV.

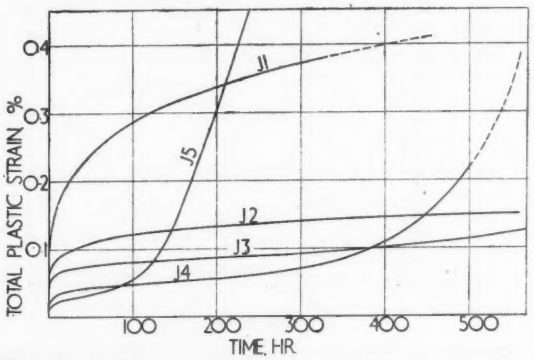


Fig. 32 (*a*)—Curves of creep tests on 0·1%-C steels with various Si contents. Stress, 8 tons/sq. in.: Temperature, 450° C.

EFFECT OF VANADIUM

In the normalized condition the effect of vanadium on molybdenum steel is greatly to increase the latent resistance to creep, but the rate at which this resistance develops tends to decrease with increasing vanadium content. Thus, even in comparatively long-time tests at a temperature of 550° C. or less, a normalized molybdenum–vanadium steel may deform to a similar or even greater extent than a molybdenum or chromium–molybdenum steel. The creep rate of the molybdenum–vanadium steel, however, continues to decrease slowly over very long periods of time long after the other steels have entered

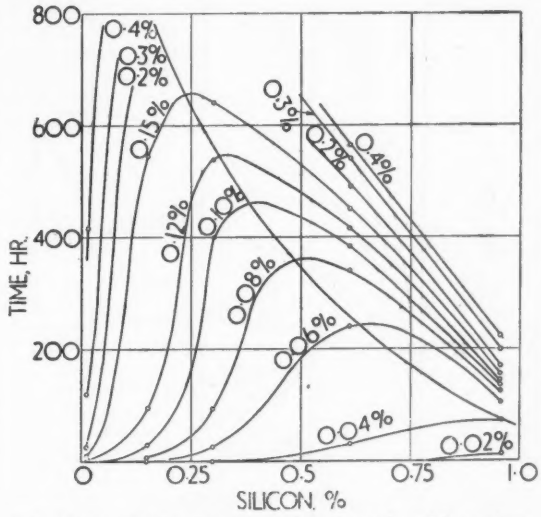


Fig. 32 (*b*)—Effect of Si in 0·1%-C steels with various Si contents, on the time to reach the plastic strains indicated. Derived from Fig. 32 (*a*). Stress, 8 tons/sq. in.: Temperature, 450° C.

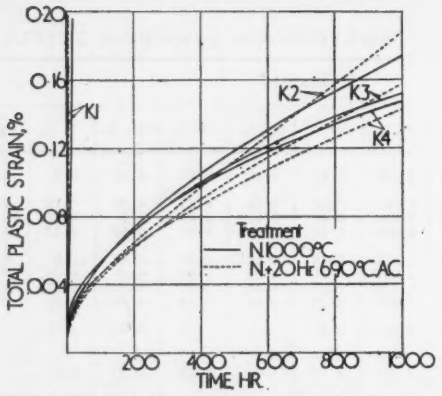


Fig. 33 (*a*)—Curves of creep tests on 0·3%-Mo/V steels (normalized at 1000° C., and normalized at 1000° C. plus 20 hr. at 690° C., A.C.). Stress, 4 tons/sq. in.: Temperature, 600° C.

the third stage of creep and have broken. Similarly it is possible to choose a molybdenum or chromium–molybdenum steel of greater short-time tensile strength than a molybdenum–vanadium steel, at a particular temperature, so that in a short-time creep test at the same temperature and with a high stress the molybdenum–vanadium steel will be inferior. However, if the comparisons are made after the steels have been given a tempering treatment the molybdenum–vanadium steel will be superior.

Creep tests at a temperature of 600° C. and a stress of 4 tons/sq. in. were carried out on two series of steels of varying vanadium content, one containing 0·3% of molybdenum and the other 0·6% of molybdenum. The steels were tested both in the normalized condition, and after normalizing and 20 hr. tempering at 690° C. (air cooled). The creep tests on the 0·3% molybdenum steels normalized at 1000° C. are shown in Fig. 33 (*a*), and the time for the specified amounts of deformation for both treatments is plotted against vanadium content in Fig. 33 (*b* and *c*).

In the normalized condition (Fig. 33 *b*) the maxima curve begins at about 0·1% of vanadium and moves to higher vanadium contents with increasing deformation. The steel with no vanadium has a creep strength of a very low order and deforms about 0·5% in 10 hr.

After tempering, the same order is maintained (Fig. 33 *c*), but the higher vanadium steels become the better in a much shorter time and the superiority of the highest vanadium steel becomes more apparent with increasing deformation. The steel with no vanadium becomes worse and deforms about 0·5% in 2 hr.

Comparing the steels with and without the tempering treatment, it is seen that although the tempering treatment improves the steels containing more than about 0·05% of vanadium in the lower range of deformation, the curves cross or tend to cross the curves of the normalized steel as the deformation increases. In this respect the highest vanadium steel is least affected.

Figures 34 (*a*, *b*, and *c*) show similar curves for the 0·6% molybdenum steels which were normalized at 950° C. In the normalized condition (Fig. 34 *b*) the maxima curve, as before, progresses to a higher vanadium content as the deformation increases. The steel with no vanadium is rather better than the 0·3% molybdenum steel and deforms 0·1% in about 120 hr.

After tempering, the steels containing about 0·07%

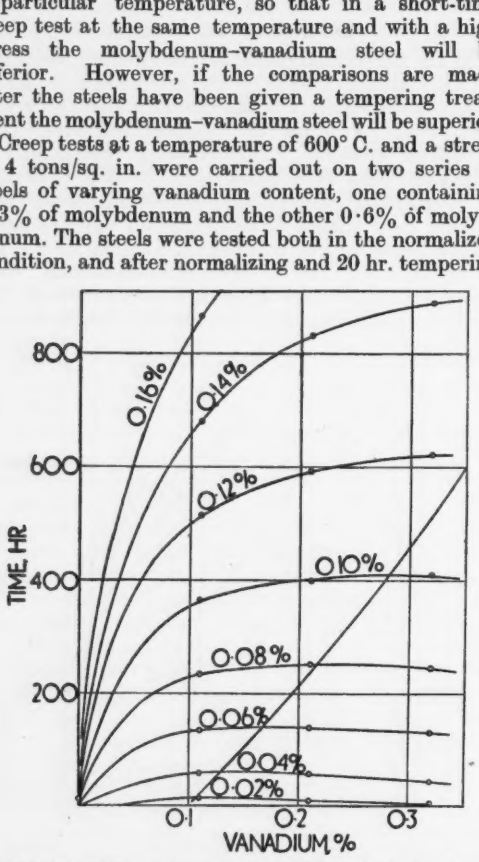


Fig. 33 (*b*)—Effect of V content of 0·3%-Mo/V steels (normalized at 1000° C.) on the time to reach the plastic strains indicated. Derived from Fig. 33 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

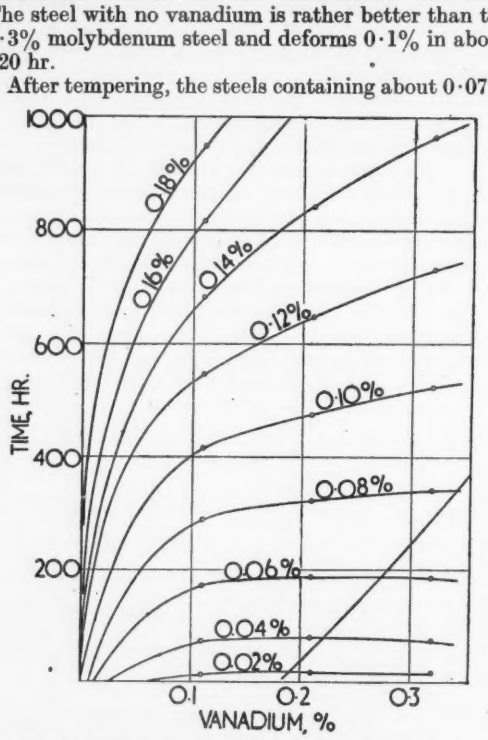


Fig. 33 (*c*)—Effect of V content of 0·3%-Mo/V steels (normalized at 1000° C. plus 20 hr. at 690° C., A.C.). Derived from Fig. 33 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

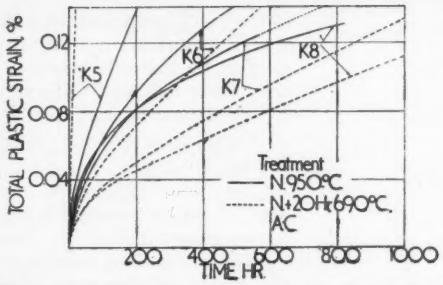


Fig. 34 (*a*)—Curves of creep tests of 0·6%-Mo/V steels (normalized at 950° C., and normalized at 950° C. plus 20 hr. at 690° C.). Stress, 4 tons/sq. in.: Temperature, 600° C.

or more of vanadium show improvement in the times considered, and if the curves of the higher vanadium steels cross those of the normalized steels this cross-over must occur after a very much longer time. The steel with no vanadium deteriorates appreciably on tempering and deforms 0·1% in 15 hr.

Comparing the data for the 0·3% and the 0·6% molybdenum steels in the normalized condition, it will be noted that the 0·3% molybdenum steels with about 0·05% or more of vanadium are better than the corresponding 0·6% molybdenum steels for quite long periods of time. The creep curve of the 0·6%-molybdenum/0·3%-vanadium steel crosses that of the corresponding 0·3% molybdenum steel after about 900 hr. and 0·15% of deformation. The 0·6% molybdenum steels with lower vanadium will cross the curves of the 0·3% molybdenum steels but only at a deformation greater than 0·15%. The above effect is partly due to the fact that the 0·3% molybdenum steels are of a lower carbon content and were normalized at a higher temperature. After tempering, the higher molybdenum steels are definitely superior.

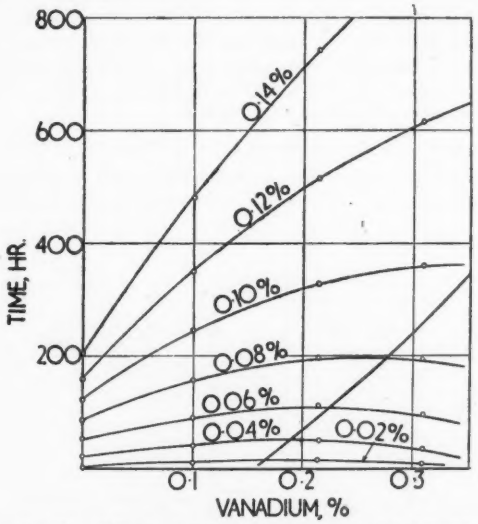


Fig. 34 (*b*)—Effect of vanadium content of 0·6%-Mo/V steels (normalized at 950° C.) on the time to reach the plastic strains indicated. Derived from Fig. 34 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

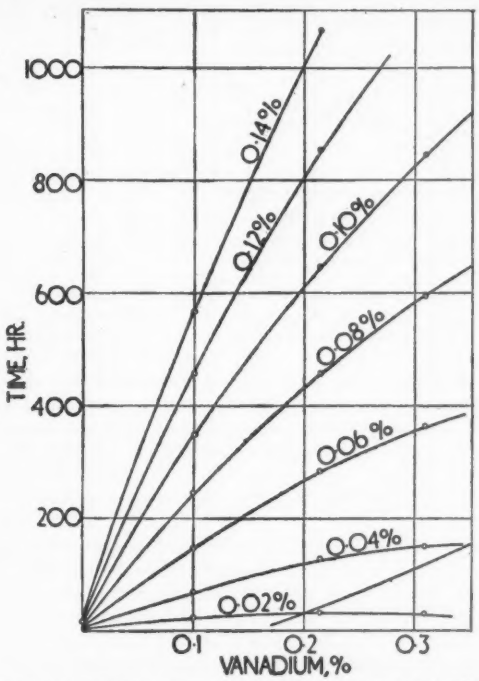


Fig. 34 (*c*)—Effect of vanadium content of 0·6%-Mo/V steels (normalized at 950° C. plus 24 hr. at 690° C.) on the time to reach the plastic strains indicated. Derived from Fig. 34 (*a*). Stress, 4 tons/sq. in.: Temperature, 600° C.

It will be realized, therefore, that in the normalized condition the low vanadium steels, particularly with low carbon and molybdenum, will be superior over much longer periods of time when tested at a lower temperature and at not too high a stress.

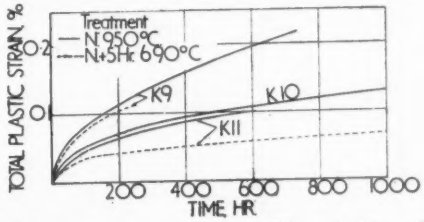
The advantage obtained by the use of vanadium is very evident from the above data, particularly because of the increased resistance to deterioration.

In the data provided below, attention has been confined to 0·5% molybdenum steel containing 0·25–0·30% of vanadium.

EFFECT OF CARBON

The tensile strength of molybdenum–vanadium steel increases very rapidly with increasing carbon, and thus for most purposes the carbon content requires to be low.

Creep tests at a temperature of 600° C. and a stress



Steel	...	...	...	K9	K10	K11
Carbon content, %			...	0·19	0·13	0·08

Fig. 35—Curves of creep tests on Mo–V steels (normalized at 950° C., and normalized at 950° C. plus 5 hr. at 690° C.) with various carbon contents. Stress, 4 tons/sq. in.: Temperature, 600° C.

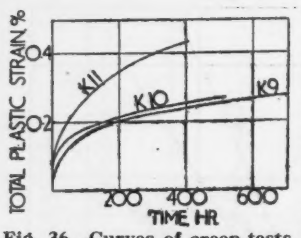


Fig. 36—Curves of creep tests on Mo–V steels (normalized at 950° C.) with various carbon contents. (For key *see* Fig. 35). Stress, 9 tons/sq. in.: Temperature, 550° C.

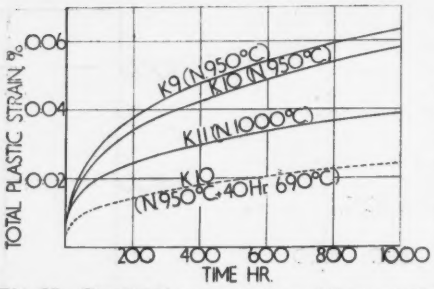


Fig. 37—Curves of creep tests on Mo–V steels with various carbon contents, after the heat-treatments indicated. Stress, 5·15 tons/sq. in.: Temperature, 538° C.

of 4 tons/sq. in. on steels of three different carbon contents are shown in Fig. 35. In the time considered, the best steel is the one with lowest carbon content. Two of the steels were tested after tempering for 20 hr. at 690° C. and it will be noted that the lower carbon steel still remains better than the higher carbon steel.

At a temperature of 550° C., and a stress of 9 tons/sq. in. the high carbon steel is the best in the time considered, whereas at a temperature of 538° C. (1000° F.) and a stress of 5·15 tons/sq. in., the low carbon steel again is the best (Figs. 36 and 37).

It is evident that carbon has a similar effect on molybdenum–vanadium steel that it has on molybdenum or chromium–molybdenum steel, and the carbon content has to be borne in mind when any comparison is being made.

EFFECT OF TEMPERING TREATMENT

The effect of tempering at 690° C. has already been illustrated in the above tests and it was seen that even this high temperature did not have the same deteriorating effect as the lower temperature (650° C.) used for the molybdenum and chromium–molybdenum steels. It was found that 50 or 100 hr. at 650° C. only indicated a gradual improvement in the creep properties and that it would require a very long time of testing to indicate whether this tempering did lead ultimately to the steel entering the third phase of creep more quickly than the normalized steel.

Creep tests at a temperature of 550° C. and a stress of 6 tons/sq. in. are shown in Fig. 38. Tempering the normalized steel improves the creep resistance considerably even when the tempering time is 1000 hr. at 650° C. or 100 hr. at 690° C. Even after 20,000 hr. of testing, the steel tempered at 690° C. for 5 hr. shows no sign of deterioration. As is seen from the the results, the 0·5% molybdenum steel is very much inferior to the molybdenum–vanadium steel.

Further tests were carried out at a temperature of 650° C. in order to indicate what the ultimate effect of tempering would be. These creep curves for a stress of 3 tons/sq. in. are shown in Fig. 39. As indicated, even after 1000 hr. at 690° C. the steel is still superior to the normalized steel up to about 0·1% of deformation, but it enters the third phase of creep more rapidly. Tempering for 1000 hr. at 650° C. also improves the steel in the initial stages, but it tends to enter the third phase of creep more quickly than the normalized steel.

The results clearly indicate that at temperatures up to at least 550° C. and for a deformation of 0·1% in 100,000 hr., molybdenum–vanadium steel is improved by tempering, even for very long periods, at a temperature as high as 690° C.

Two other curves are shown in Fig. 39, and one of them refers to a creep test on the open-hearth molybdenum–vanadium steel which had been furnace-cooled from 950° C. This treatment seriously reduces the creep strength. The other curve (of steel K10) shows that an electric-furnace cast of steel is not so good as an open-hearth cast when normalized at the same temperature. The steels used were of similar composition.

EFFECT OF NORMALIZING TEMPERATURE

It was shown for molybdenum and chromium–molybdenum steels that increasing the normalizing

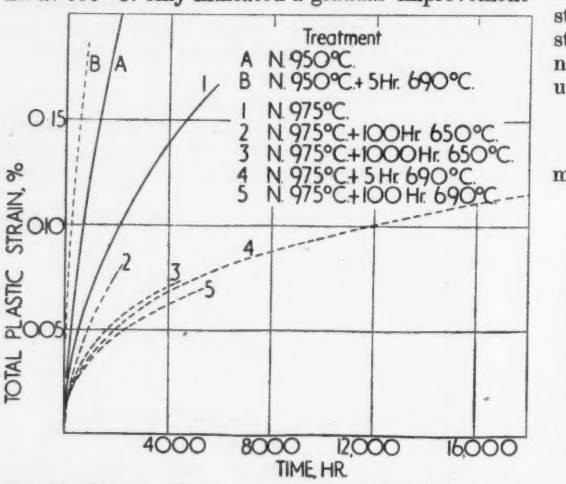


Fig. 38—Curves of creep tests on 0·09%-C/0·5%-Mo steel A2 and Mo–V steel K12 after heat-treatments indicated (steel A2, curves A and B: steel K12, curves 1–5). Stress, 6 tons/sq. in.: Temperature, 550° C.

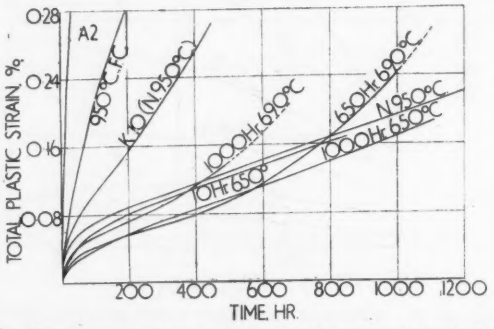


Fig. 39—Curves of creep tests on Mo–V steel K12 after heat-treatments indicated. Stress, 3 tons/sq. in.: Temperature, 650° C.

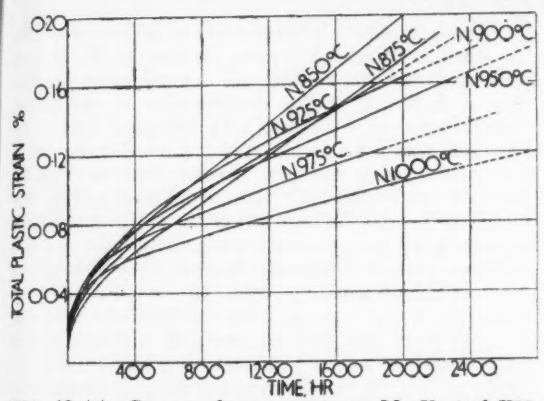


Fig. 40 (*a*)—Curves of creep tests on Mo–V steel K12 normalized at temperatures indicated. Stress, 6 tons/sq. in.: Temperature, 550° C.

temperature first of all caused a decrease and then an increase in the deformation in the initial stages of the test, but with longer times the curves switched over until the test at the highest normalizing temperature became the best. A similar effect occurs with molybdenum–vanadium steel.

Creep curves at a temperature of 550° C. and a stress of 6 tons/sq. in. are shown in Fig. 40 (*a*), the tests being carried on for approximately 2000 hr.

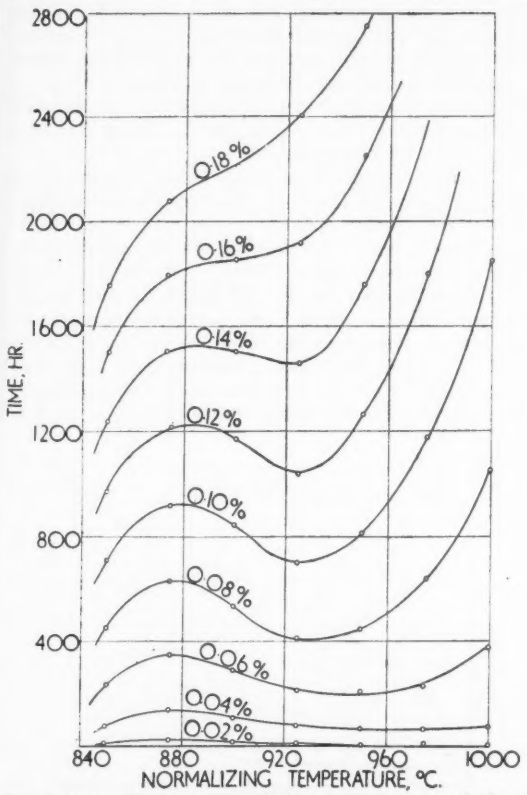


Fig. 40 (*b*)—Effect of normalizing temperature of Mo–V steel K12 on the time to reach the plastic strains indicated. Derived from Fig. 40 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

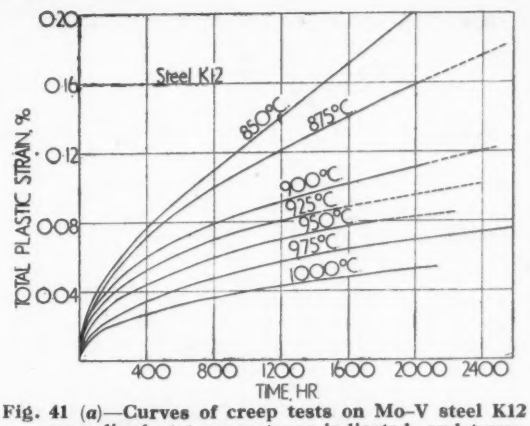


Fig. 41 (*a*)—Curves of creep tests on Mo–V steel K12 normalized at temperatures indicated, and tempered for 5 hr. at 690° C. Stress, 6 tons/sq. in.: Temperature, 550° C.

on a basic open-hearth steel. The change-over in the curves is indicated more clearly by plotting the normalizing temperature against time for specified deformation, as shown in Fig. 40 (*b*). It requires about 2000 hr. of test before the steels line themselves up in order of normalizing temperature, the highest temperature giving the best result.

Similar tests were also carried out after tempering for 5 hr. at 690° C. and air cooling. In this condition, the steels align themselves in order of the normalizing temperature from the start of the test. The curves are shown in Figs. 41 (*a* and *b*). The noticeable features about these results are the great advantage obtained by increasing the normalizing temperature

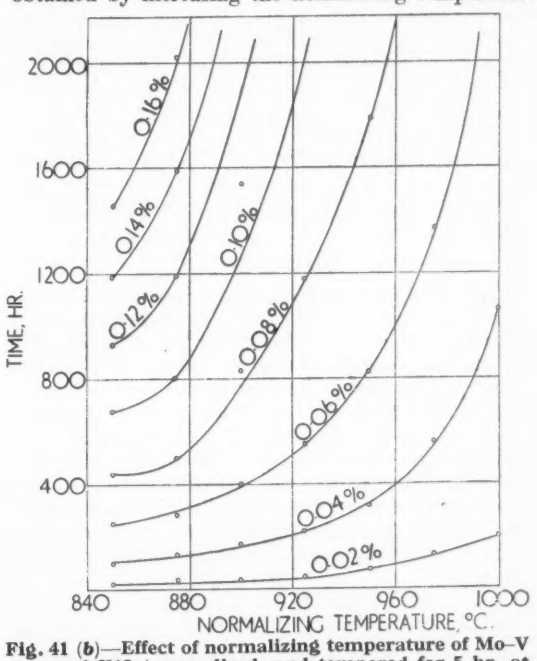


Fig. 41 (*b*)—Effect of normalizing temperature of Mo–V steel K12 (normalized, and tempered for 5 hr. at 690° C.) on the time to reach the plastic strains indicated. Derived from Fig. 41 (*a*). Stress, 6 tons/sq. in.: Temperature, 550° C.

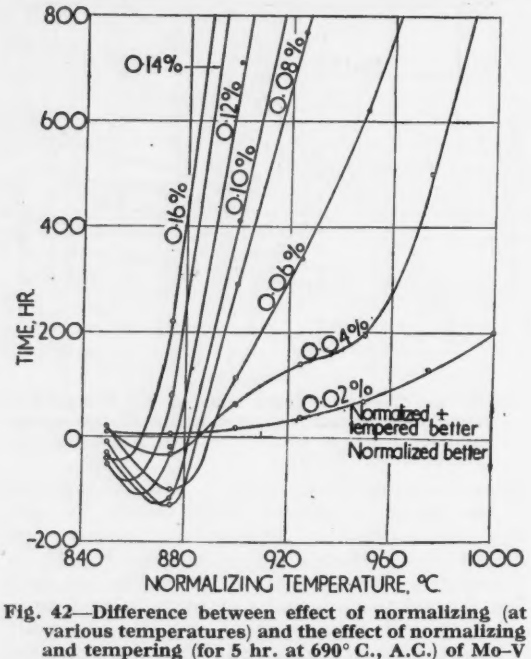


Fig. 42—Difference between effect of normalizing (at various temperatures) and the effect of normalizing and tempering (for 5 hr. at 690° C., A.C.) of Mo–V steel K12 on the time to reach the plastic strains indicated. Derived from Figs. 40 (*b*) and 41 (*b*). Stress, 6 tons/sq. in.: Temperature, 550° C.

and secondly, the steepness of the curves in Fig. 41 (*b*) with increasing normalizing temperature. Thus, it is difficult to duplicate creep-test results in such a steel even with very careful control of the heat-treatment.

In Fig. 42, the difference in time for specific deformation taken from the curves in Fig. 40 (*a* and *b*) has been plotted. With the lower normalizing temperature, the tempering treatment appears to be slightly detrimental when considering a deformation below about 0·16%. Even in the worst case, however, there is no appreciable deterioration and the test indicates that the steel would be better at still longer times.

DISCUSSION

The foregoing tests all indicate that molybdenum–vanadium steel has essentially the same characteristics as that of molybdenum or chromium–molybdenum steel, but that these characteristics are much more pronounced. In particular, molybdenum–vanadium steel is very much superior to the other types in that at normal working temperatures the steel continues to stiffen for an exceedingly long time and that deterioration will not occur in the normal life-time of the steel. Indeed, molybdenum–vanadium steel should be tempered before going into service in order to obtain the full advantage of this stiffening effect.

RUPTURE TESTS

Rupture tests must be carried out at higher stresses and/or temperatures than creep tests if failure is to be obtained in a reasonable time. When the stress is high, the results will tend to favour the steels of highest short-time tensile strength and steels which give the best results in creep tests at high stress. As the stress is lowered, however, the time to fracture is prolonged and the results will depend more on the rate of stiffening and/or deterioration of the steel. Consideration of creep-test data indicates that very long times may be necessary for this effect to come into play. Creep tests are usually continued up to deformations of not more than 0·5%, and within this range the increase in true stress due to the reduction in area of the test-piece will not greatly influence the results. With rupture tests, however, the true stress may differ appreciably from the nominal stress, depending on the creep characteristics of the steel. The effect of this on two steels of different creep characteristics will depend on the time taken to rupture. If rupture occurs in a relatively short time this effect will obviously favour the steel with superior initial resistance. On the other hand, if the time to rupture is sufficiently prolonged the effect will favour the steel of superior latent resistance.

When rupture tests are carried out at higher temperatures than are normal for creep tests, oxidation of the surface of the test-piece is liable to be severe with a consequent increase in true stress. It is only to be expected, therefore, that the size of the test-piece will have a considerable influence on the time to rupture, the smaller the diameter of the test-piece the shorter the time, since for any given amount of scaling, the percentage reduction in area is greater the smaller the original diameter of the test-piece. Another effect of oxidation is accelerated intercrystalline failure due to preferential oxidation at deformed and stressed grain boundaries. This is small at temperatures below about 550° C. Even *in vacuo* failure may be intercrystalline.

A further complication of great importance is the mode of failure of the steel. In a tensile test at high temperature, failure usually occurs with considerable elongation and a large reduction of area, and similarly with short-time rupture tests. In longer times ductility tends to decrease and failure is initiated by intercrystalline cracking. At constant temperature and with increasing time to fracture, the ductility reaches a minimum after a certain time and then with longer time it increases. With decreasing temperature this minimum value decreases, and in

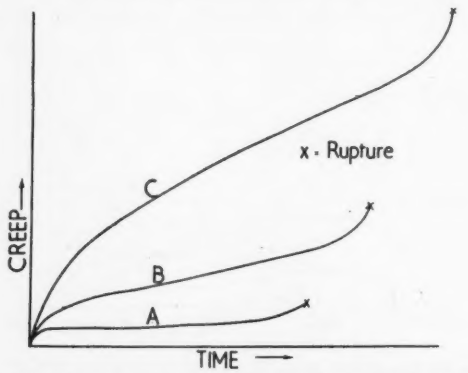


Fig. 43—Curves of creep tests continued to fracture, illustrating that rupture strength may be inverse to creep resistance

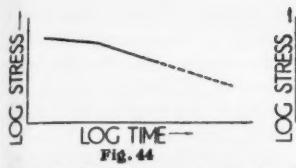


Fig. 44

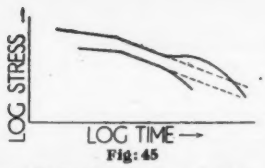


Fig. 45

Fig. 44—Log.-stress-to-fracture/log.-time curve at constant temperature, illustrating the method of extrapolation to give stress causing fracture in 100,000 hr.

Fig. 45—Log.-stress-to-fracture/log.-time curves, illustrating deviation of actual results from extrapolated line

extreme cases failure may occur with only about 1% of elongation. This has been clearly demonstrated by Tapsell and others[12] for a molybdenum and molybdenum–vanadium steel, the method used being to plot elongation contour curves showing the effect of temperature and time.

Thus, although it may appear that rupture tests are simpler to interpret than creep tests because there is only one criterion, *i.e.*, the time to fracture, in fact, the problem is even more complicated than in creep testing. If failure always occurred with large elongation, then rupture tests would simply be the end points of creep tests, and creep properties and rupture properties would be directly related. Since elongations at failure may vary considerably, it may happen that, in a comparison between several steels, the steel of best creep properties has the worst rupture strength. This is illustrated in Fig. 43. Though in general the deformation at fracture decreases with increasing creep resistance, it cannot be concluded that the steel of best creep resistance will show the least deformation at fracture.

The first comprehensive study of steel by means of rupture tests was carried out in the United States of America by White, Clark, and Wilson[13, 14] on carbon and carbon–molybdenum steels in the annealed condition. It was found that when the logarithm of the stress was plotted against the logarithm of the time to fracture, the results could be represented by two straight lines, as shown in Fig. 44. The point of intersection of these lines was said to represent the onset of intercrystalline failure.

By carrying out a series of tests at different temperatures and by extrapolating straight lines from a few thousand hours to 100,000 hr., the rupture-strength/temperature curve for this time is obtained, but such results are open to several serious objections. At temperatures where scaling is important the log./log. curve is not straight but curves downward. Since practically all of the published work has been carried out on annealed steels, it is to be expected that with these steels in the normalized condition the enhanced stiffening effect of the latter treatment would result in improved rupture strength at an early stage. Later, when deterioration sets in, the curve would tend downwards again. These effects are illustrated in Fig. 45.

Tapsell[12] obtained the long-time rupture strength of a molybdenum and a molybdenum–vanadium steel by a method suggested by Bailey.[15] Tests were carried out at several stresses and various temperatures, and a family of constant-stress curves were drawn plotting the temperature against the logarithm of the time to fracture. As with the American data, the curves obtained were essentially straight lines over a considerable part of the length, and were extrapolated as straight lines. In any method, the influence of scaling, stiffening, and deterioration must be taken into account by carrying out tests of sufficient duration, so that the true trend of the curves is obtained. The tests of the molybdenum–vanadium steel carried out by Tapsell were insufficiently long to obtain a true estimate of its capabilities. In fact, the straight-line

Table V

RESULTS OF RUPTURE TESTS ON MOLYBDENUM STEELS

O.H. = Open-hearth furnace, E.F. = Electric furnace, H.F. = High-frequency furnace. Elongations at fracture given in brackets

Steel	Type	Alloying Elements		Heat-Treatment	Temperature of Test, °C.	Time to Fracture at Stress Shown, hr.									
		C, %	Mo, %			12 tons/sq. in.	10 tons/sq. in.	9 tons/sq. in.	8 tons/sq. in.	7 tons/sq. in.	6 tons/sq. in.	5 tons/sq. in.	4 tons/sq. in.	3 tons/sq. in.	2·3 tons/sq. in.
A2	O.H.	0·09	0·51	Normalized at 950° C.	650	..	..	..	..	..	10¼ (30)	30 (28)	78 (30)	297 (15)	767 (25)
A9	E.F.	0·13	0·55	,, ,, 950° C.	650	..	..	..	..	..	10 (34)	..	93 (27)	348 (18)	516 (28)
A4	O.H.	0·21	0·55	,, ,, 950° C.	650	..	..	..	..	16 (20)	..	101* (14)	276* (11)	660 (15)	1162† (18)
A8	H.F.	0·12	1·33	,, ,, 950° C.	650	..	..	..	..	15 (9)	..	104 (7)	325 (5)	1828 (6)	..
A2	O.H.	0·09	0·51	,, ,, 950° C.	600	..	85 (18)	..	172 (15)	..	430 (8)	..	1820 (14)	3850 (18)	..
A8	H.F.	0·12	1·33	,, ,, 950° C.	600	113 (10)	175 (4)	..	320 (2)	930 (2)	2102 (2)	..	..	..	..
A2	O.H.	0·09	0·51	,, ,, 950° C.	550	..	..	1500 (5)	2152 (4)	3203 (5)	5502 (8)	14200 (10)	..	..	..
A2	O.H.	0·09	0·51	,, ,,1050° C.	550	..	..	..	1480 (3)	2220 (4)	..	..	..	..	..

* 0·357-in. dia. test-piece † 0·504-in. dia. test-piece

Table VI

RESULTS OF RUPTURE TEST ON CHROMIUM–MOLYBDENUM STEELS

O.H. = Open-hearth furnace, E.F. = Electric furnace, H.F. = High-frequency furnace. Elongations at fracture given in brackets

Steel	Type	Alloying Elements				Heat-Treatment	Temperature of Test, °C.	Time to Fracture at Stress Shown, hr.											
		C, %	Si, %	Mo, %	Cr, %			12 tons/sq. in.	10 tons/sq. in.	9 tons/sq. in.	8 tons/sq. in.	7 tons/sq. in.	6 tons/sq. in.	5 tons/sq. in.	4 tons/sq. in.	3 tons/sq. in.	2·3 tons/sq. in.
B1	O.H.	0·10	0·14	0·53	0·72	Normalized at 950° C.	650	..	..	..	..	74 (33)	..	193 (30)	445 (25)	1280 (26)	2550 (25)
B3	H.F.	0·22	0·23	0·57	0·48	,, ,, 950° C.	650	..	..	..	..	17 (30)	38 (33)	..	204 (20)	720 (22)	2155 (20)
H1	H.F.	0·07	0·06	0·62	0·72	,, ,, 950° C.	650	..	..	..	..	..	21 (24)	..	144 (22)	461 (25)	1476 (25)
H2	H.F.	0·07	0·23	0·62	0·75	,, ,, 950° C.	650	..	..	..	..	..	14 (35)	..	85 (40)	392 (29)	1112 (16)
H3	H.F.	0·065	0·38	0·54	0·77	,, ,, 950° C.	650	..	..	..	..	..	9 (38)	..	79 (35)	301 (28)	1108 (23)
B2	O.H.	0·145	0·15	0·54	0·75	,, ,, 875° C.	650	..	..	..	..	..	169 (28)	125 (20)	206 (28)	..	..
B2	O.H.	0·145	0·15	0·54	0·75	,, ,, 975° C.	650	..	..	..	..	..	98 (35)	162 (22)	336 (25)	..	..
B2	O.H.	0·145	0·15	0·54	0·75	,, ,, 1000° C.	650	..	..	..	..	..	110 (33)	183 (29)	412 (15)	..	..
B1	O.H.	0·10	0·14	0·53	0·72	,, ,, 950° C. + 5 hr. at 690° C.	650	..	..	..	..	..	13 (40)	27 (35)	91 (33)	182 (28)	..
B1	O.H.	0·10	,,	,,	,,	Normalized at 950° C.	600	120 (20)	205 (22)	..	395 (18)	630 (12)	1164 (12)	..	5320 (14)	12560 (16)	..
B2	O.H.	0·145	0·15	0·54	0·75	,, ,, 950° C.	600	..	..	..	592 (16)	..	1720 (14)	..	5645 (15)	..	..
B1	O.H.	0·10	0·14	0·53	0·72	,, ,, 950° C.	550	..	..	5664 (9)	6286 (9)	..	9794 (14)	16000 (U)	..	..	..
B2	O.H.	0·145	0·15	0·54	0·75	,, ,, 950° C.	550	..	..	..	6705 (10)	..	18200 (12)	..	58000 (U)	..	..

(U) = Unbroken

extrapolation gave a low result because of the change in shape of the curves due to the considerable stiffening which occurred in the tests at very high temperatures.

In the original planning of the rupture tests reported below, the object was to carry out comparative tests on various steels, and a temperature of 650° C. was chosen in order to accelerate the time to fracture. It was found that the order of merit depended on the temperature and stress and it was realized that very long time tests at lower temperatures were therefore necessary. Part of the equipment was set aside for tests at temperatures of 600° and 550° C. and stresses of 4, 5, and 6 tons/sq. in. Further tests at these temperatures were put in hand later when the position became clearer due to the accumulation of creep and rupture-test data. Sufficient tests have now been completed to indicate the relative merits of the three types of steel under consideration.

PROCEDURE

The test-pieces used were of 0·252 in. diameter (0·05 sq. in. in area) with a parallel gauge length of 2 in. A few tests were, however, carried out with test-pieces of 0·357 in. diameter (0·1 sq. in. in area) and 0·504 in. diameter (0·2 sq. in. in area). The elongation was measured between pop marks on the shoulders of the test-pieces, the increase in length being taken to be the increase on the 2 in. parallel gauge length.

To avoid confusion in the graphical presentation of the results, the data from which the curves were obtained are not indicated on the graphs but are shown in Tables V, VI, and VII. Since none of the tests was of sufficiently short duration to give transcrystalline fractures, no indication of the two straight lines effect shown in Fig. 44 was obtained. When the results were plotted on the log.-stress/

Table VII

RESULTS OF RUPTURE TESTS ON MOLYBDENUM–VANADIUM STEELS

O.H. = Open-hearth furnace, E.F. = Electric Furnace, H.F. = High-frequency furnace. Elongations at fracture given in brackets

Steel	Type	Alloying Elements			Heat-Treatment	Temperature of Test, °C.	Diameter of Test-piece, in.	Time to Fracture at Stresses Shown, hr.											
		C, %	Mo, %	Va, %				14	12	11	10	9	8	7	6	5	4	3	2·3	1·7
K12	O.H.	0·12	0·54	0·265	Normalized at 975° C.	690	0·252	:	:	:	:	:	:	:	38 (15)	110 (18)	295 (20)	540 (25)	720 (28)	807 (26)
K12	O.H.	,,	,,	,,	,, ,, 975° C.	650	0·252	:	:	128 (7)	170 (8)	:	293 (4)	396 (8)	630 (5)	1290 (5)	1490 (6)	2880 (8)	3706 (8)	:
K12	O.H.	,,	,,	,,	,, ,, 975° C.	650	0·357	:	:	:	:	:	:	:	855 (9)	1520 (14)	:	3520 (10)	:	:
K12	O.H.	,,	,,	,,	,, ,, 975° C.	650	0·504	:	:	:	:	:	:	:	:	:	2430 (10)	3910 (11)	4650 (12)	:
K10	E.F.	0·13	0·62	0·31	,, ,, 950° C.	650	0·252	:	:	15 (18)	14 (14)	36 (10)	60 (9)	:	260 (8)	:	1116 (10)	2350 (10)	3594 (11)	:
K 9	E.F.	0·19	0·53	0·31	,, ,, 950° C.	650	0·252	:	:	:	65 (8)	:	115 (5)	:	232 (3)	:	1104 (4)	:	3036 (4)	:
K12	O.H.	0·12	0·54	0·265	Normalized at 975° C. + 5 hr. at 690° C.	650	0·357	:	:	32 (18)	48 (19)	68 (16)	110 (18)	230 (14)	672 (11)	1920 (12)	2184 (12)	4760 (14)	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 5 hr. at 690° C.	650	0·504	:	:	:	:	:	:	:	:	:	3136 (14)	:	7000 (U)	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 40 hr. at 690° C.	650	0·252	:	:	:	:	:	:	:	330 (13)	:	1571 (18)	:	2784 (14)	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 40 hr. at 690° C.	650	0·357	:	:	:	29 (22)	54 (18)	98 (16)	260 (17)	650 (14)	:	1871 (15)	:	3096 (18)	4716 (14)
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 1000 hr. at 690° C.	650	0·252	:	:	:	1070 (6)	1560 (6)	2840 (4)	:	10820 (4)	21010 (5)	144 (26)	390 (26)	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C.	600	0·252	:	530 (5)	:	660 (9)	:	2280 (10)	:	15840 (9)	:	:	1102 (24)	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 5 hr. at 690° C.	600	0·252	:	182 (15)	:	195 (12)	:	780 (12)	:	5100 (14)	:	44000 (U)	:	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 40 hr. at 690° C.	600	0·252	:	85 (20)	:	10880 (3)	:	:	:	:	:	:	:	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C.	550	0·252	:	5905 (5)	6300 (2)	8600 (11)	15600 (9)	16000 (U)	16000 (U)	58000 (U)	:	50000 (U)	:	:	:
K12	O.H.	,,	,,	,,	Normalized at 975° C. + 5 hr. at 690° C.	550	0·252	1240 (11)	3005 (12)	:	:	:	:	:	:	:	:	:	:	:

(U) = Unbroken

log.-time-to-fracture scale, it was found that at the higher stresses they fell on a straight line. This line was extrapolated in order to serve as a base line to indicate stiffening or deterioration, as shown in Fig. 45.

However, a more convenient way of presenting the results is to plot stress against log. time-to-fracture. In this way the effect of stress is more clearly seen, and any change in the direction of the curves is more noticeable. Advantage was still taken of the log./log. method of plotting by transferring the extrapolated straight line on to the stress/log.-time-to-fracture curves. In these graphs the extrapolated part of the line is shown by a hatched curve. For simplicity this will be referred to as the log./log. line. Extrapolation of the actual experimental curve is shown by a dotted line.

RESULTS

Rupture Tests at 650° C.

Molybdenum Steels (*Fig. 46*)—All the curves with one exception follow the log./log. line for varying lengths of time and then gradually fall below it. The exception is the steel containing 1·33% of molybdenum (curve 5) which follows the log./log. line up to the limit of time tested.

The deterioration of the steels, *i.e.*, the drop below the

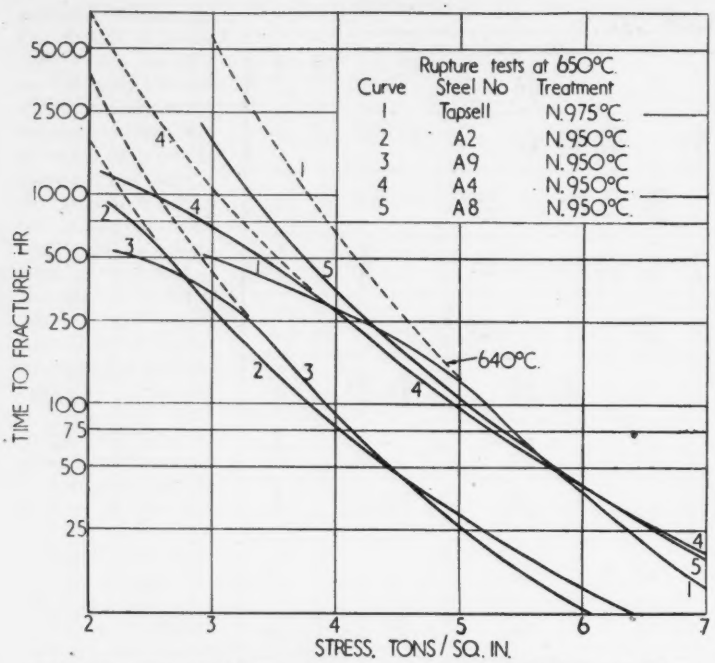


Fig. 46—Stress/log.-time-to-fracture curves at 650° C. of Mo steels normalized at indicated temperatures

log./log. line, occurs after a minimum of 100 hr. Thus apparently rupture tests at 650° C. are a less sensitive criterion of deterioration than is the effect of tempering at 650° C. on creep tests at a lower temperature. It will be noted later, however, that tempering affects rupture curves in a rather different manner.

Since scaling is severe at 650° C. it might be thought that this explains the drop below the log./log. line. Scaling has some effect but it cannot be the full explanation, since the drop occurs at times varying from 100 to more than 2000 hr. in these steels which all have similar scaling properties.

Similarly, intercrystalline cracking is not the full explanation, since the steels which are less ductile (particularly the high molybdenum steel) continue on the log./log. line for a greater length of time.

The controlling factor, therefore, must be the spheroidization of the steel, and the behaviour of the steels in the rupture tests should be in the same order as their behaviour in creep tests. Thus, at high stresses the 0·21% carbon open-hearth steel (curve 4) is better than the 0·09% carbon steel (curve 2), but as the stress is decreased the curve of the 0·21% carbon steel approaches that of the low carbon steel and would probably cross it at a still lower stress. Also, the electric-furnace steel (curve 3) deteriorates more rapidly than the open-hearth steel (curve 2). These effects are in agreement with the behaviour in creep testing.

A curve (1) is also shown for a 0·09% - carbon/0·5% - molybdenum steel, the data being taken from Tapsell's work.[12] The temperature of testing, however, was 10° lower. This steel had a very fine structure and also contained 0·27% of chromium. Due to this chromium content and because of the slightly lower temperature of testing, it is better than the 0·09% carbon open-hearth steel (curve 2) at the higher stresses. Because of its fine structure, however, it deteriorates more rapidly than the open-hearth steel.

Chromium – Molybdenum Steels (*Fig. 47*)—It will be noted that the curves agree well with the log./log. lines. The low carbon electric-furnace steels drop below the line, this effect occurring more quickly with increasing silicon content.

The open-hearth steels gave the best results, particularly the 0·1% carbon steel (curve 1) which was much superior to the corresponding 0·5% molybdenum steel.

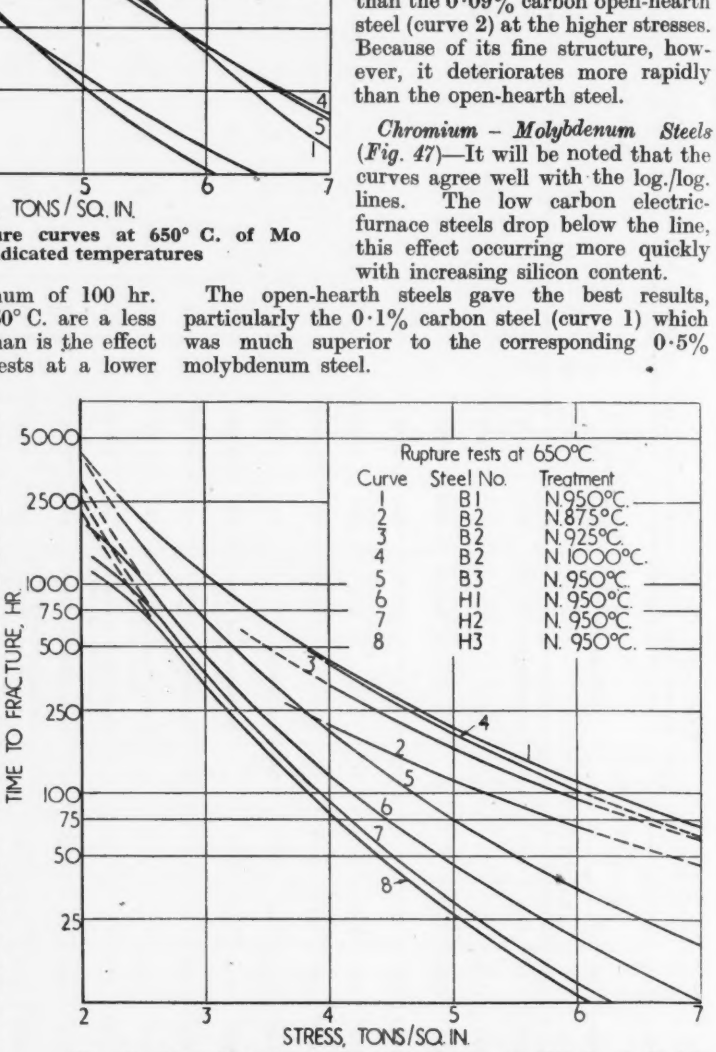


Fig. 47—Stress/log.-time-to-fracture curves at 650° C. of Cr–Mo steels normalized at indicated temperatures

The effect of varying the normalizing temperature is also shown for the chromium–molybdenum steel containing 0·145% of carbon, and the results indicate that with increasing normalizing temperature the rupture strength increases.

Molybdenum – Vanadium Steels (*Fig. 48*)—The most notable feature of these curves is the marked rise above the log./log. line which occurs after a definite time. This effect can be attributed only to the stiffening effect which occurs in these steels.

The electric-furnace steel (curve 3), although of very similar composition to the open-hearth steel (curve 2), showed inferior properties. The high carbon electric-furnace cast (curve 4) was superior to the lower carbon steel (curve 3) at high stresses, but was inferior at stresses below 6 tons/sq. in. This is in agreement with the conclusions based on creep tests regarding the effect of the carbon content.

A curve (1) is also shown for Tapsell's molybdenum – vanadium steel tested at 655° C. The data were not sufficient to determine the log./log. line, but it is obvious that

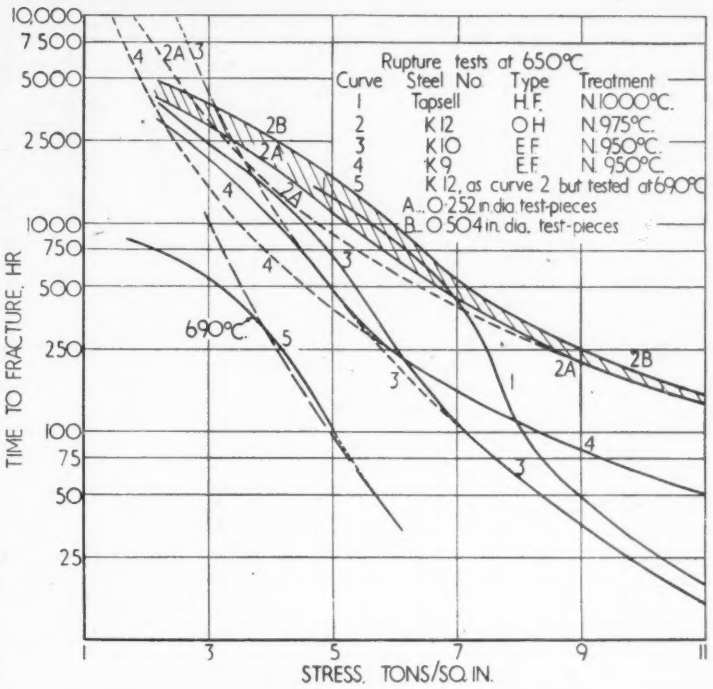

Rupture tests at 650°C
Curve Steel No. Type Treatment
1 Tapsell H.F. N.1000°C.
2 K12 O.H. N.975°C.
3 K10 E.F. N.950°C.
4 K9 E.F. N.950°C.
5 K12, as curve 2 but tested at 690°C
A...0·252 in. dia. test-pieces
B...0·504 in. dia. test-pieces
690°C.
TIME TO FRACTURE, HR.
STRESS, TONS/SQ. IN.

Fig. 48—Stress/log.-time-to-fracture curves at 650° C. of Mo–V steels normalized at indicated temperatures

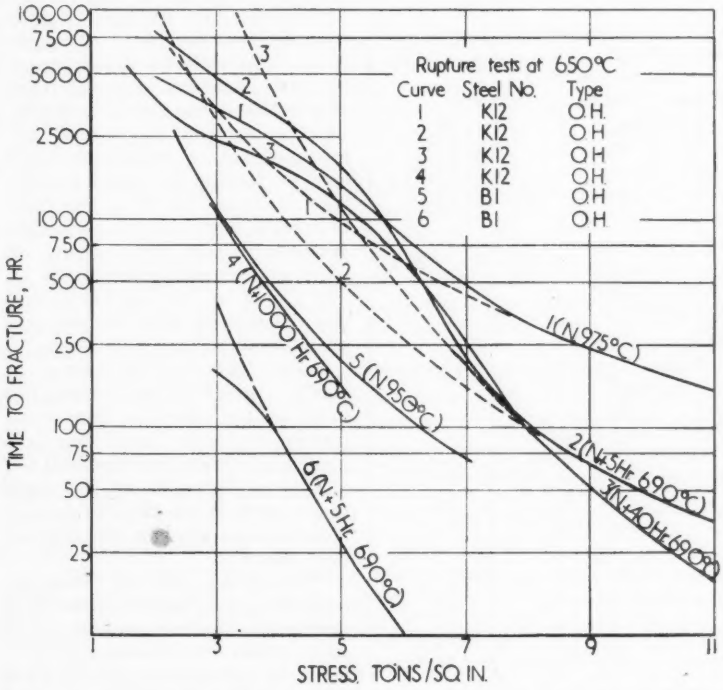


Fig. 49—Stress/log.-time-to-fracture curves at 650° C. of Mo–V and Cr–Mo steels at heat-treatments indicated

marked stiffening also occurs with this steel.

A curve (5) for the open-hearth steel tested at 690° C. is also shown on this graph (Fig. 48). The stiffening effect was slight and the curve then dropped below the log./log. line.

Effect of Test-Piece Size—The open-hearth molybdenum–vanadium steel was tested at 650° C. using three different sizes of test-piece (Table VII). With increasing test-piece diameter, a longer time to fracture was obtained. For this reason the results have been represented by a band, as shown in Fig. 48. As already discussed, this effect should be less the shorter the time to fracture and/or the lower the testing temperature.

The molybdenum–vanadium steels are very much superior to either the molybdenum or chromium–molybdenum steels in tests at 650° C. In fact the molybdenum–vanadium steel tested at 690° C. is superior to the low carbon 0·5% molybdenum steel tested at 650° C.

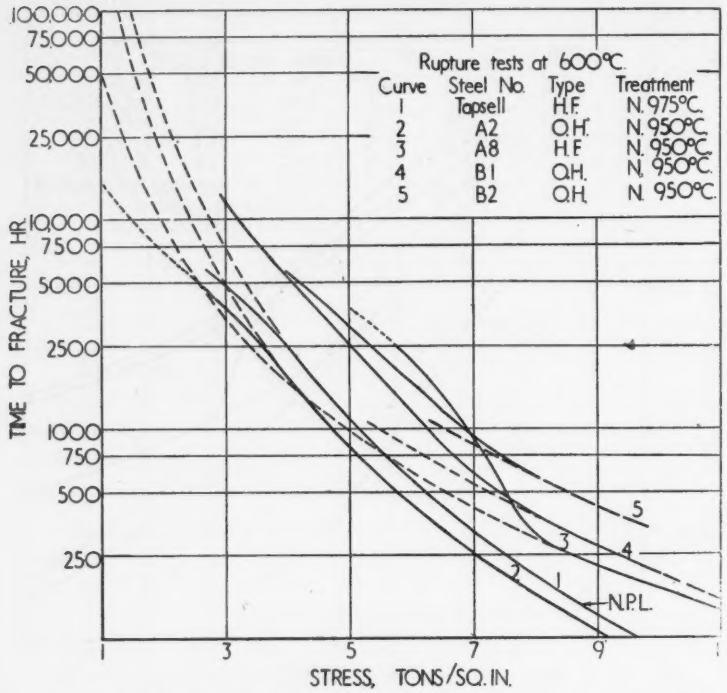


Fig. 50—Stress/log.-time-to-fracture (at 600° C.) curves of Mo and Cr–Mo steels normalized as indicated

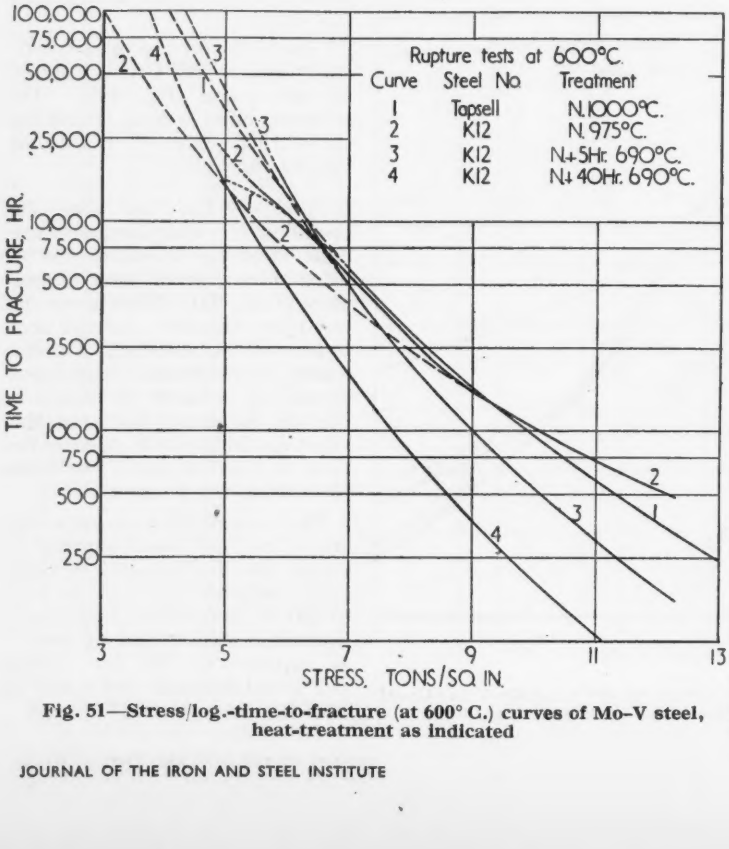


Fig. 51—Stress/log.-time-to-fracture (at 600° C.) curves of Mo–V steel, heat-treatment as indicated

Effect of Tempering (*Fig. 49*)—The effect of tempering was investigated mainly for the open-hearth molybdenum–vanadium steel, although a few results were obtained on chromium–molybdenum steel. The curves represent results obtained using 0·357 in. dia. test-pieces.

Due to tempering, the rupture strength at high stresses is decreased, but at the same time tempering alters the slope of the log./log. line so that as the stress is lowered the log./log. line of the tempered steel approaches and may even cross that of the normalized steel. However, the actual curves may not continue to follow the log./log. line, since the additive effect of tempering due to the time at the testing temperature must also be considered.

With the molybdenum–vanadium steel, stiffening takes place in tests at 650° C. even when the steel was tempered for 40 hr. at 690° C. and, due to the greater slope of the log./log. line in this condition, this stiffening is sufficient to enable the curve of 40 hr. tempering (3) to cross that of the 5-hr. tempering treatment (2), but with 5 hr. tempering a more marked stiffening occurs so that the 5-hr. curve again crosses that of the 40-hr. tempering. Similarly, the 5-hr. tempering curve crosses that of the normalized steel, but the effect of time at 650° C. is not sufficient to enable the latter steel to develop its maximum rupture strength, so that the 5-hr. tempering curve remains superior. However, if tests were carried out at a higher temperature, say at 700° C., it is very probable that the normalized steel in this case would be superior at all stages.

With 1000 hr. tempering at 690° C. sufficient data were not obtained to define the curve completely, but even after this prolonged treatment the steel still has a rupture strength of the same order as that of the chromium–molybdenum steel in the normalized condition.

The log./log. curve of the tempered chromium–molybdenum steel approaches that of the normalized steel as the stress decreases, but the effect of testing time at 650° C. is such that the curve of the tempered steel falls below that of the log./log. line, since the tempered steel has

little capacity for further stiffening. At a rather lower temperature, however, a tempered chromium–molybdenum steel would be better at low stresses than the normalized steel.

Rupture Tests at 600° C.

Molybdenum Steels (*Fig. 50*)—The 0·09% carbon steel (curve 2) follows the same trend as Tapsell's steel (curve 1) but with a slightly lower rupture strength. No stiffening occurs, and the steel drops below the log./log. line after about 2500 hr. of test. The 1·3% molybdenum steel (curve 3) which showed no deterioration in tests at 650° C. shows pronounced stiffening at 600° C.

The following results indicate the effect of tempering at 650° C. on the rupture strength at 6 tons/sq. in. and a temperature of 600° C.:

Treatment	Time to Fracture 0·09% Carbon Steel	Time to Fracture 0·21% Carbon Steel
Normalized at 950° C. ...	430 hr.	1015 hr.
Normalized + 5 hr. at 650° C.	321 hr.	348 hr.

This tempering treatment markedly reduced the time to rupture particularly in the case of the 0·21% carbon steel.

Chromium–Molybdenum Steel (*Fig. 50*)—Curves are shown for the 0·145% carbon (5) and the 0·10% carbon (4) open-hearth steels. The higher carbon steel is better at the higher stresses but tends to become inferior to the lower carbon steel as the stress is reduced. Both steels show considerable stiffening.

Molybdenum–Vanadium Steel (*Fig. 51*)—The open-hearth molybdenum–vanadium steel was tested in the normalized condition (curve 2) and after 5 and 40 hr. tempering at 690° C. (curves 3 and 4, respectively). As in the rupture tests at 650° C., tempering reduced the rupture strength at the higher stresses, but at lower stresses the tempered steel tends to become the better. The curve of the steel tempered for 5 hr. crosses that of the normalized steel despite the more marked stiffening in the latter case.

Tapsell's results on molybdenum–vanadium steel are also shown in Fig. 51 (curve 1). When reproduced in this way, the extrapolated part of the curve indicates deterioration of the steel. Tapsell's own results at 655° C. indicate stiffening, and thus it would appear that Tapsell's extrapolation underestimates the strength of the steel at 600° C.

Rupture Tests at 550° C.

Molybdenum Steels (*Fig. 52*)—The 0·09% carbon open-hearth molybdenum steel (curve 2) begins to stiffen up after about 5000 hr. of testing. The curve of this steel has been extrapolated to 100,000 hr. as indicated. Tapsell's steel (curve 1) follows the log./log. line up to 20,000 hr. but the extrapolation is such that the curve then falls below this line, although at this temperature some stiffening is to be expected in this steel.

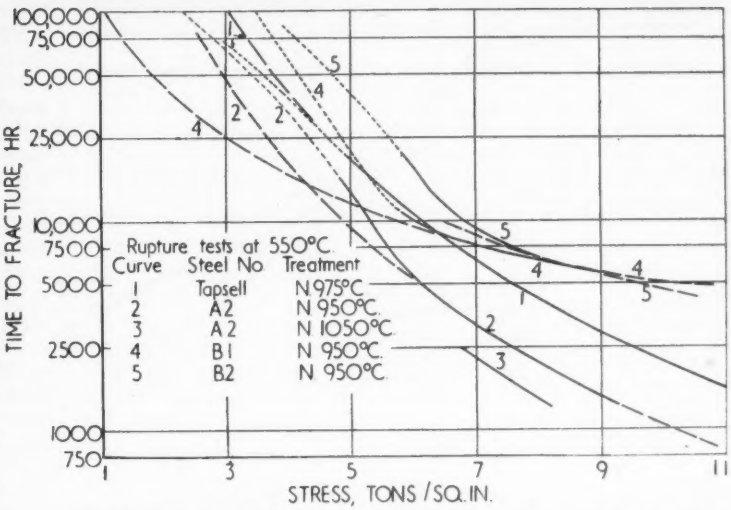


Fig. 52—Stress/log.-time-to-fracture (at 550° C.) curves of Mo and Cr–Mo steels normalized as indicated

The open-hearth steel was also tested after coarsening the structure by a normalizing treatment at 1050° C. This has the effect of lowering the rupture strength, probably due to premature failure accompanied by low elongation.

Chromium–Molybdenum Steel (*Fig. 52*)—Curves are also shown for the 0·145% carbon (5) and the 0·10% carbon (4) chromium–molybdenum steels. As at 600° C., pronounced stiffening occurs, and the curves have been extrapolated to 100,000 hr. This extrapolation is justified by the test on the 0·145% chromium–molybdenum steel which is still unbroken after 58,000 hr. at 4 tons/sq. in. Since the 0·1% carbon steel showed greater stiffening in creep tests, it is to be expected that its rupture strength will tend to become better than that of the 0·145% carbon steel at lower stresses.

Molybdenum–Vanadium Steel (*Fig. 53*)—At 550° C. only three tests were made on the open-hearth steel in the normalized condition (curve 2) and these results indicate that the steel has superior rupture strength to the tempered steel at high stresses, but the log./log. lines cross at about 15,000 hr. of test. More stiffening, however, should occur with the normalized steel in longer times, but this will take place very slowly at this temperature. The curve (3) of the tempered steel has been extrapolated to 100,000 hr. This extrapolation is justified by the test-piece still unbroken after 58,000 hr. at 6 tons/sq. in. The extension of this test-piece to date is only about 0·2%.

Tapsell's data on molybdenum–vanadium steel is also shown (curve 1) and indicates a marked deterioration of the steel in long tests. Such deterioration does not occur with molybdenum–vanadium steels and thus it must be concluded that Tapsell's extrapolation underestimates the rupture strength of molybdenum–vanadium steel at this temperature.

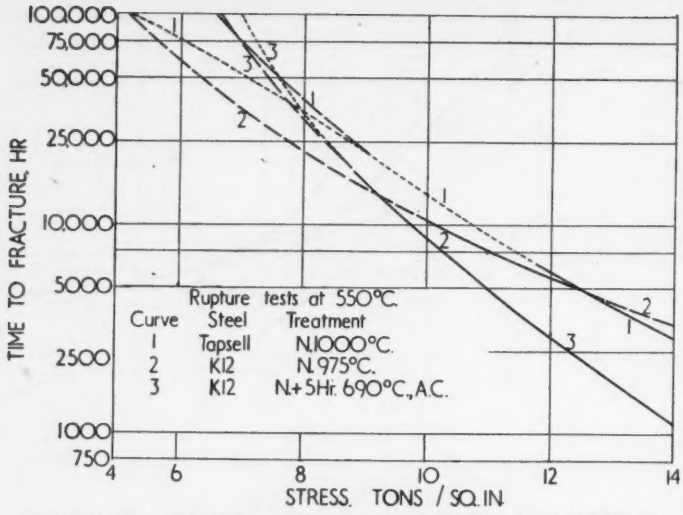


Fig. 53—Stress/log-time-to-fracture (at 550° C.) curves of Mo–V steels, heat-treatment as indicated

will show up to still greater advantage.

Molybdenum–vanadium steel in the normalized and tempered condition is much superior to the other steels. Tapsell's molybdenum–vanadium steel is similar to the tempered steel at high stresses but is apparently inferior at lower stresses. It has already been demonstrated, however, that this is an underestimate for molybdenum–vanadium steel in the normalized condition.

In Fig. 56 Tapsell's curves at 7 tons/sq. in. have been reproduced. This line has also been reproduced in Fig. 54. As shown in this latter figure the slope of this line is greater than that of the molybdenum steel, which would apparently indicate that at still lower temperatures the molybdenum steel would become the better. This conclusion is not justified, so that the curve of the normalized steel must bend over in the manner indicated in Fig. 56.

Rupture-Stress/Temperature Curves

The stress/time curves of the open-hearth 0·09%-carbon/0·5%-molybdenum steel, the 0·1%-carbon/0·5%-molybdenum/0·8%-chromium steel, and the 0·5%-molybdenum/0·25%-vanadium steel were used to estimate the rupture-stress/temperature curves in the manner reported by Tapsell. The temperature/time curves at constant stress for the above three steels are shown in Figs. 54, 55, and 56. Although rupture tests were carried out only at three temperatures it is considered that the extrapolations are reasonably accurate, since the results fall very nearly on straight lines. If anything, this method of extrapolation will tend to underestimate the strength of the steels. As was demonstrated above, this method of extrapolation is not applicable to molybdenum–vanadium steel in the normalized condition. The rupture-strength/temperature curves for a time of 100,000 hr. are shown in Fig. 57. Tapsell's results[12] on molybdenum and normalized molybdenum–vanadium steels have been included.

As is shown, the molybdenum steel is inferior to Tapsell's steel, particularly at high stresses. This is accounted for by the fact that Tapsell's steel contained 0·66% of manganese, compared with 0·43% for the open-hearth steel and 0·27% of chromium compared with 0·03% for the open-hearth steel.

The chromium–molybdenum steel is slightly better than the 0·5% molybdenum steel, particularly at high stresses. It must be emphasized, however, that the chromium–molybdenum steel stiffens to a greater degree and is less prone to deterioration than the molybdenum steel. Thus, in practice, where conditions of heat-treatment are not ideal, chromium–molybdenum steel

Ductility

In any high-temperature installation, local stresses may be present which raise the actual stress at certain points to values above that of the designed stress. Such stresses are relieved by creep, and thus it may be dangerous to use a steel which fails with a very small deformation due to intercrystalline cracking, even although the creep strength of the steel is very good.

Tapsell has shown that 0·5% molybdenum steel

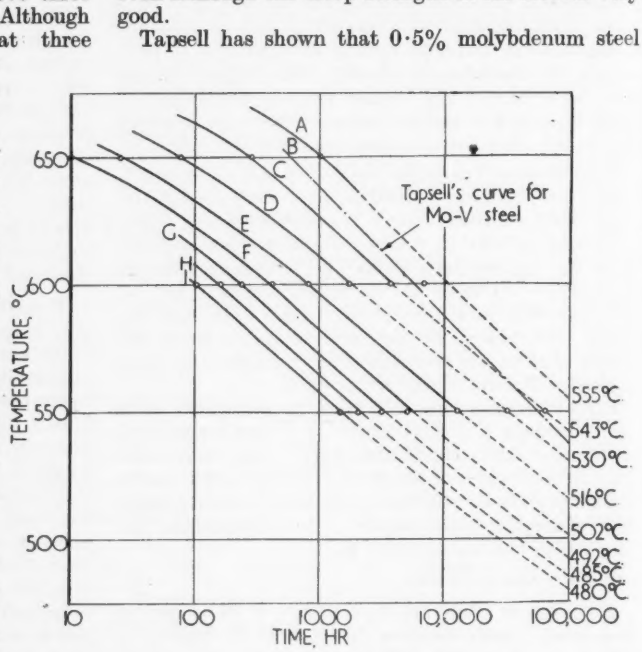
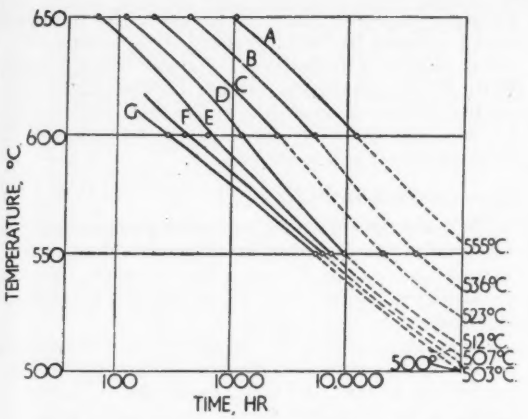


Fig. 54—Temperature/log.-time-to-fracture curves of 0·09%-C/0·5%-Mo steel A2 (normalized at 950° C.) for various stresses



Curve	A	B	C	D	E	F	G
Stress, tons/sq. in. ...	3	4	5	6	7	8	9

Fig. 55—Temperature/log.-time-to-fracture curves of 0·1%-C/0·8%-Cr/0·5%-Mo steel B1 (normalized at 950° C.) for various stresses

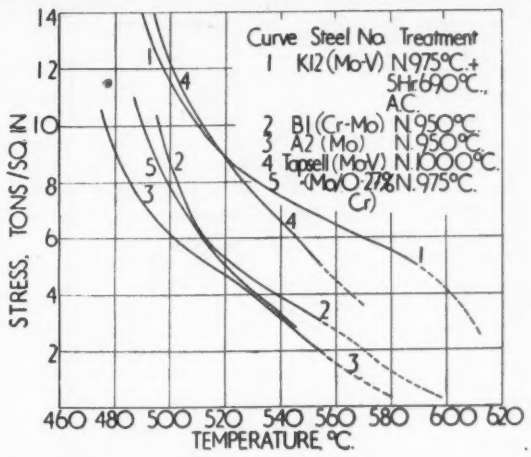


Fig. 57—Stress/temperature curves of Mo, Cr–Mo, and Mo–V steels for rupture in 100,000 hr. Heat-treatment as indicated

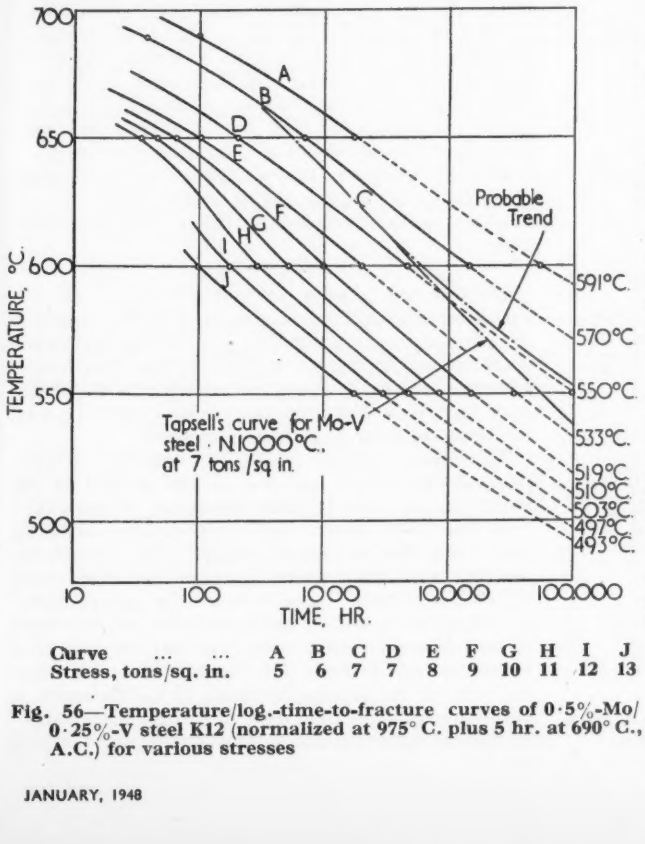
may fail in an intercrystalline manner with very low elongation, and that molybdenum–vanadium steel in the normalized condition is even more brittle.[12] This is illustrated by the elongation contour curves in Figs. 58 and 59 reproduced from his paper.

Molybdenum Steels—In Fig. 58 the elongation figures for the rupture tests of the 0·09% carbon open-hearth steel have been inserted. The low elongation values are in good agreement with Tapsell's results, but the higher values tend to be less. This is, however, to be expected, since the above tests were measured on a 2-in. gauge length although the area was only 0·05 sq. in., whereas Tapsell's figures were measured on $4\sqrt{\text{Area}}$. When the elongation is low there is no appreciable necking of the test-pieces, so that the length of the gauge length has little influence. The results confirm Tapsell's conclusion that 0·5% molybdenum steel may fail with very low elongation.

From the tabulated results (Table V) it will be noted that the higher carbon 0·5% molybdenum steel gives slightly lower elongation values than the low carbon steel. The 1·33% molybdenum steel is very brittle, particularly at 600° C., and presumably it will be even more brittle at lower temperatures. A very coarse structure decreases the ductility, as shown by the results on the 0·09% carbon steel normalized at 1050° C.

Chromium–Molybdenum Steel—The results on chromium–molybdenum steel (Table VI), show the same tendency as those of molybdenum steel, but the minimum values of elongation obtained are of the order of 10%, despite the better creep and rupture strength of the chromium–molybdenum steel.

Molybdenum–Vanadium Steel—In Fig. 59 the elongation values shown in Table VII, of the tests on molybdenum–vanadium steel in the normalized and in the normalized plus 5 hr. at 690° C. conditions, have been inserted. The values for the normalized steel confirm Tapsell's results and show that molybdenum–vanadium steel in the normalized condition may fail with a very low elongation.



Curve	A	B	C	D	E	F	G	H	I	J
Stress, tons/sq. in.	5	6	7	7	8	9	10	11	12	13

Fig. 56—Temperature/log.-time-to-fracture curves of 0·5%-Mo/0·25%-V steel K12 (normalized at 975° C. plus 5 hr. at 690° C., A.C.) for various stresses

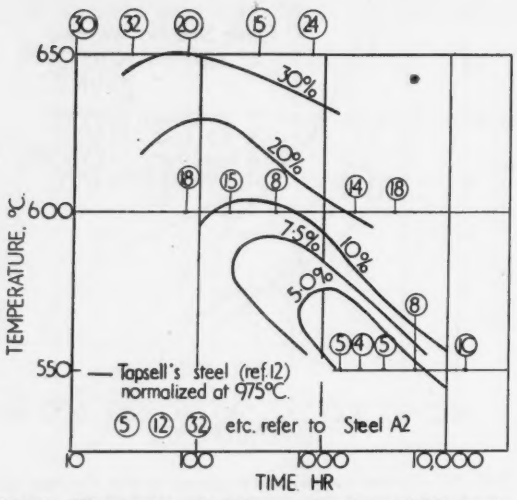


Fig. 58—Effect of varying time to fracture of Mo steels on the elongation at various temperatures

It is well known that tempering increases the ductility of steels at high temperatures, but in general this also results in a decrease in creep and rupture strength properties. Molybdenum–vanadium steel is an exception to this rule, since tempering for 5 hr. at 690° C. improves the long-time creep and rupture properties as well as the ductility. However, at high stresses, *i.e.*, short times to fracture, the tempered steel is inferior to the normalized steel, so that there is some degree of similarity of behaviour to other steels. With 40 hr. tempering at 690° C. the ductility is further increased, as shown in the tabulated results. In this case, however, the rupture strength will be reduced except at low stresses.

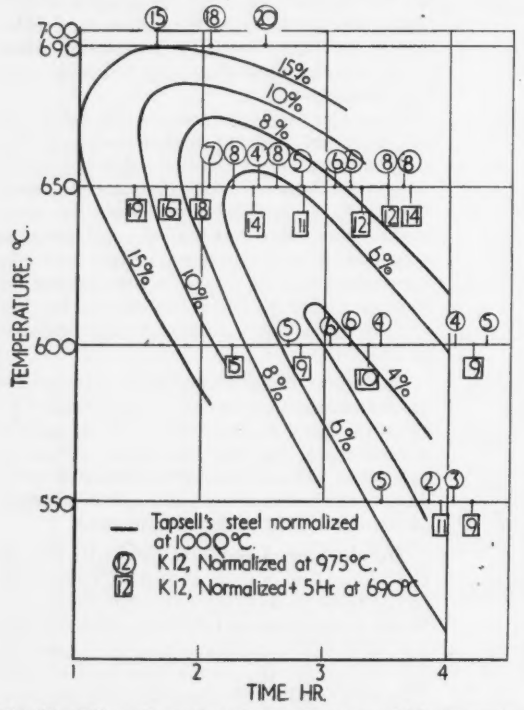


Fig. 59—Effect of varying time to fracture of Mo–V steels on the elongation at various temperatures

Intercrystalline Cracking

The reduction of area gives a truer measure of the degree of strain preceding fracture than does elongation, because even though the test-piece size is standardized, steels vary in the amount of uniform deformation before necking takes place.

Owing to the scaling it was not possible to measure accurately the reduction of area in the rupture tests carried out at 600° and 650° C. At 550° C., however, scaling is so negligible as not to interfere with the measurement of reduction in area. The test results obtained at 550° C. and shown in Tables V, VI, and VII are reproduced in Table VIII, with the addition of reduction of area.

Taking reduction of area as a criterion of ductility, the steels fall in the same order as when compared on the basis of elongation, but the tempered molybdenum–vanadium steel and the chromium–molybdenum steel show up to still greater advantage.

Examination of the fractures showed that all of the failures were intercrystalline, even for the chromium–molybdenum steel which showed 80% reduction of area. Another notable feature was that the cracking in the case of the chromium–molybdenum and the tempered molybdenum–vanadium steel occurred only in the necked portion of the test-piece. It must be concluded, therefore, that intercrystalline cracking can occur only after a certain minimum deformation, which minimum, depending on the type of steel and on its heat-treatment, varies over a wide range. Reduction of area gives a more exact comparison of this deformation than does elongation.

Further work on the nature of intercrystalline cracking is in progress and it is intended to give a more extensive discussion on this subject at a later date.

SUMMARY

Rupture-test results are influenced by all the variables which influence creep tests and in addition are affected by other variables such as intercrystalline cracking and variation in true stress. To ensure accurate assessment of rupture strength at 100,000 hr. very long time tests are necessary, exceeding even the times used for long-time creep tests.

To ensure reproducibility of rupture results, particularly at high stresses, careful control of the variables, particularly heat-treatment, is essential.

It has been shown that molybdenum–vanadium steel has a greater rupture strength than chromium–molybdenum steel which in turn is better than molybdenum steel. Tapsell's values for molybdenum–vanadium steel are an underestimate, because tests of sufficient duration were not carried out, the maximum time being of the order of 5000 hr. Tempering reduces the rupture strength of molybdenum or chromium–molybdenum steel, but with molybdenum–vanadium steel it reduces the strength only at high

stresses, whereas the tempered steel may be even better than the normalized steel at lower stresses.

Molybdenum steel is very prone to failure with low elongation, whereas chromium–molybdenum steel fails with good ductility although the fracture is still intercrystalline. Molybdenum–vanadium steel in the tempered condition also shows good ductility, but in the normalized condition its ductility is even worse than that of the 0·5% molybdenum steel.

STRESS/TEMPERATURE RELATIONSHIP FOR DEFORMATION IN 100,000 HR.

It is evident from the preceding discussion that the long-time creep properties of a steel expressed, for example, as the stress/temperature relationship for 0·1% of deformation in 100,000 hr., require a considerable amount of long-time creep testing, and that even when this has been carried out the results can be applied strictly only to the particular sample of steel tested. It is therefore necessary to employ some form of short-time test to ensure that supplies are up to the standard of the prototype.

Sufficient data have been collected for carbon and 0·5% molybdenum steels to enable the stress/temperature relationship to be estimated as shown in Figs. 2 (*a* and *b*).

Similar data have not been published for chromium–molybdenum steel, but there is no doubt that this type of steel has better long-time creep properties than 0·5% molybdenum steel. It is probable, as was shown in the case of rupture tests, that the difference in long-time creep properties between the two steels will not be appreciable under laboratory conditions of testing, but in practice chromium–molybdenum steel shows up to advantage since it is less adversely influenced by variations in heat-treatment and is less subject to deterioration. Another and more important feature of chromium–molybdenum steel is that, unlike 0·5% molybdenum steel, it possesses adequate ductility in long-time creep and rupture tests.

Tapsell[12] has estimated the stress/temperature relationship for 0·2% of deformation in 100,000 hr. for normalized molybdenum–vanadium steel as is shown in Fig. 64. As he himself indicates, this curve was based on somewhat crude measurements of extension during tests to rupture. Since these tests were of comparatively short duration (the longest test taking approximately 1000 hr. to reach 0·2% of deformation) and were carried out at high stresses, the results can be considered as only approximate.

Tests lasting up to 4500 hr., of much longer duration than those of Tapsell's tests, have been carried out on this type of steel in the normalized condition and an estimate of the stress/temperature relationship made by the method used and described by Tapsell.[5, 6]

MOLYBDENUM–VANADIUM STEEL NORMALIZED

The steel used was made in an electric furnace, normalized at 950° C., and tested at different temperatures under stresses of 6 and 4 tons/sq. in. The results of the creep tests plotted on a log.-deformation/log.-time scale are shown in Figs. 60 and 61. These curves were extrapolated in a manner considered to show the probable trend, although the shape of the curves is such that rather longer times of tests at the lower temperatures would be necessary to obtain very accurate extrapolation.

It will be noted that pronounced stiffening occurs during the creep tests, this stiffening taking place at a later stage the lower the temperature of the test, *i.e.*, it is due to the development of latent resistance to creep caused by the tempering effect of time at the testing temperature. It will also be noted that the stiffening becomes slightly less pronounced the lower the testing temperature. This indicates that molybdenum–vanadium steel does not develop its maximum creep resistance when in the normalized condition except at high temperatures of testing.

Data from the curves were used to construct temperature/log.-time curves of iso-strain, as shown

Table VIII

COMPARISON OF REDUCTIONS IN AREA FOR VARIOUS STEELS

Steel	Heat-Treatment	Property	Stress, tons/sq. in.							
			14	12	10	9	8	7	6	5
0·5% Mo steel (A2)	Normalized at 950° C.	Time to fracture, hr.	...	...	...	1500	2152	3203	5502	14,200
		Elongation, %	...	...	...	5	4	5	8	10
		Reduction of area, %	...	...	...	17	14	12	16	24
Cr–Mo steel (B1)	Normalized at 950° C.	Time to fracture, hr.	...	...	...	5664	6286	...	9794	...
		Elongation, %	...	...	...	9	9	...	14	...
		Reduction of area, %	...	...	...	46	52	...	80	...
Mo–V steel (K12)	Normalized at 975° C.	Time to fracture, hr.	...	5905	10,880	...	...	...	...	...
		Elongation, %	...	5	3	...	...	...	...	...
		Reduction of area, %	...	6	5	...	...	...	...	...
Mo–V steel (K12)	Normalized at 975° C. + 5 hr. at 690° C.	Time to fracture, hr.	1240	3005	8600	15,600	...	...	...	...
		Elongation, %	12	12	11	9	...	...	...	...
		Reduction of area, %	41	44	34	36	...	...	...	..

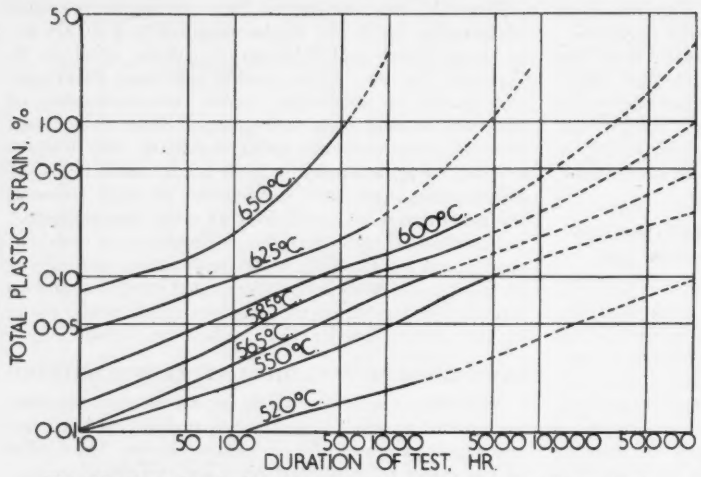


Fig. 60—Log.-strain/log.-time curves of tests on Mo–V steel K10 (normalized at 950° C.) at various temperatures indicated. Stress, 4 tons/sq. in.

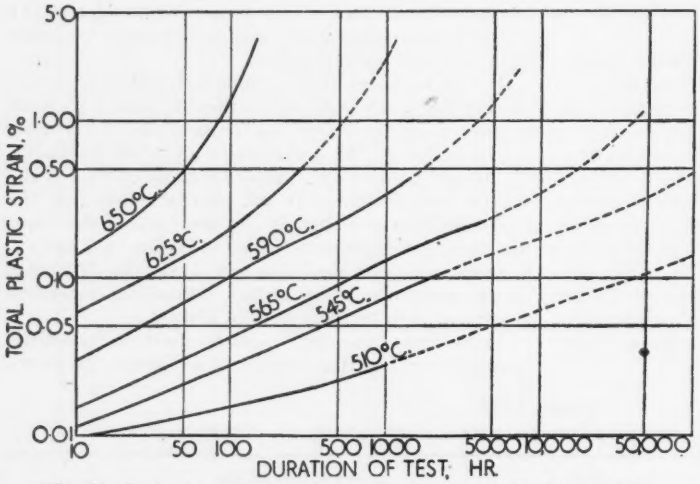


Fig. 61—Log.-strain/log.-time curves of tests on Mo–V steel K10 (normalized at 950° C.) at various temperatures indicated. Stress, 6 tons/sq. in.

in Figs. 62 and 63. At high temperatures the curves tend to converge to zero time, and at low temperatures they tend to converge to infinite time, *i.e.*, the curves are of reversed **S** shape. This is the fundamental shape of such curves, and is particularly noticeable when the steel has a pronounced latent resistance to creep. With 0·5% molybdenum steel and with carbon steels the reversed **S** shape is not pronounced.

As has already been shown in the discussion of rupture tests, a steel with pronounced latent resistance to creep must be investigated with tests of sufficient duration to bring out this effect, particularly at the lower temperatures, otherwise the tendency is to extrapolate the curves in a way which underestimates the long-time creep properties of the steel. This is illustrated by the dotted curves in Fig. 62, which ignore the tests at 550° C. and 520° C.

In Fig. 64 the stress/temperature curves for 0·1 and 0·2% of deformation in 100,000 hr. for the normalized molybdenum–vanadium steel are shown and compared with Tapsell's values for 0·2% of creep.[12]

The open-hearth molybdenum–vanadium steel will have rather better long-time creep properties than the electric-furnace steel reported above, but the general shape and trend of the curves would be the same.

MOLYBDENUM–VANADIUM STEEL IN THE NORMALIZED AND TEMPERED CONDITION

During the course of the foregoing tests and the short-time and rupture tests already described, it was realized that molybdenum–vanadium steel only develops its best creep properties when in the normalized and tempered condition. Accordingly, long-time creep tests were carried out on the open-hearth molybdenum–vanadium steel which had been normalized at 975° C. and then tempered for 5 hr. at 690° C. and air cooled. The tests were carried out at stresses of 6, 5, and 4 tons/sq. in. These were sufficient to provide data in the working range of stress and temperature. At stresses lower than 4 tons/sq. in. exceedingly long times of test at high temperatures would be necessary to establish the trend of the curves. It was not considered necessary to do tests at higher stresses, and indeed this was not possible for the volume of other work would have been severely restricted. The tests at 5 tons/sq. in. were not so extensive as those at 6 and 4 tons/sq. in. and are not reported in detail.

Figures 65 and 66 show the creep curves plotted on the log./log. scale basis and it will be noted that at a stress of 6 tons/sq. in. and a temperature of 550° C. the creep test was carried out for 20,000 hr. This curve confirms the small deformation of the rupture test-pieces at the same temperature and stress at which they are still unbroken after 58,000 hr. of test. The marked stiffening observed for the normalized steel was not present, but stiffening does occur though in a more gradual manner.

Temperature/log.-time curves of iso-strain were obtained as before, and these curves are shown in Figs. 67 and 68. Comparison of these Figs. with Figs. 62 and 63 for the normalized electric-furnace steel brings out several notable features which would have been even more striking if the comparison had been made with the open-hearth steel in the normalized condition.

In the first place, the reversed **S** shape of the curves of the tempered steel is much less pronounced than that of the normalized steel, and the curves of the tempered steel thus approximate in shape to those obtained by Tapsell for carbon steel[5] and 0·5% molybdenum steel.[6]

Secondly, it will be noted that at a stress of 6 tons/sq. in., the normalized steel has a better creep strength at 650° C. than has the tempered steel, although at lower temperatures the tempered steel is by far the better, *i.e.*, the effect of tempering the molybdenum–vanadium steel is to reduce the creep strength at high temperatures but to improve it at lower temperatures. The higher the stress the lower is the temperature at which the normalized steel remains the better, as is confirmed by the fact that at a temperature of 650° C. and a stress of 4 tons/sq. in. the tempered steel still remains the better. This effect is the same as that which occurs in rupture tests, *i.e.*, at any temperature the normalized steel is the better at high stresses but the tempered steel tends to become the better, or becomes the better, as the stress is lowered.

The above effect is a general one and is not restricted to molybdenum–vanadium steel, but it is only in more complex steels that it assumes practical importance. With carbon steels, 0·5% molybdenum steels, or chromium–molybdenum steels the range of temperature in which the effect of tempering is beneficial is too low for it to be of any practical significance. Moreover, for these steels the extent of the possible improvement is small.

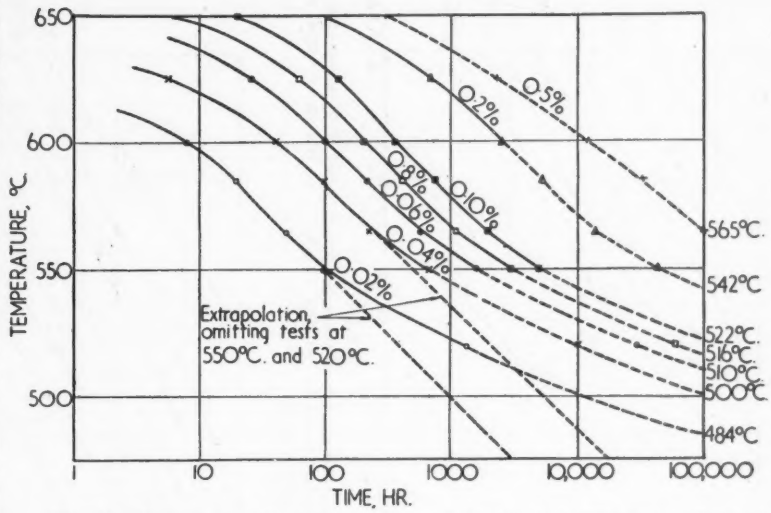


Fig. 62—Temperature/log.-time curves of Mo–V steel K10 (normalized at 950° C.) at plastic strains indicated. Stress, 4 tons/sq. in.

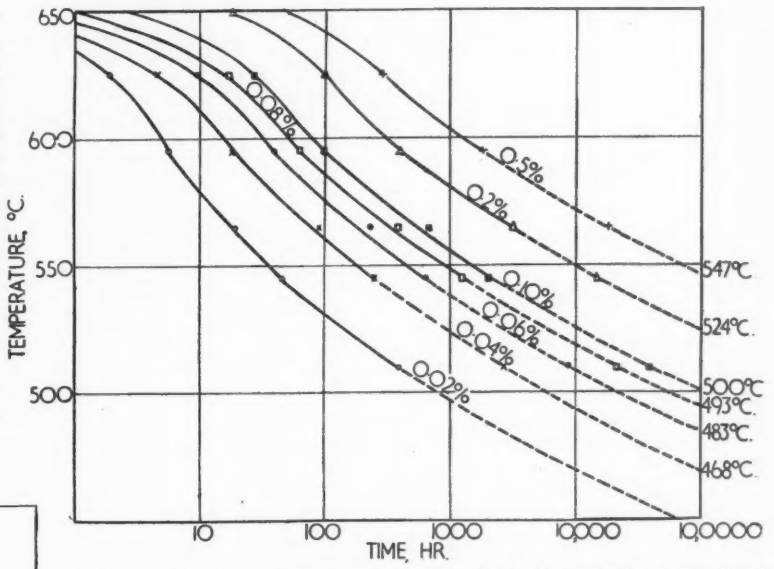


Fig. 63—Temperature/log.-time curves of Mo–V steel K10 (normalized at 950° C.) at plastic strains indicated. Stress, 6 tons/sq. in.

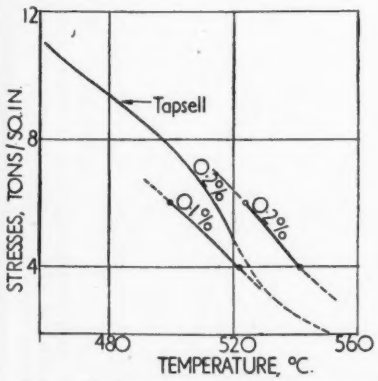


Fig. 64—Stress/temperature curves of Mo–V steel (normalized) for 0·1% and 0·2% of plastic strain in 100,000 hr.

In Figs. 69 (*a* and *b*) the stress/temperature curves for 0·1 and 0·2% of creep are shown for the tempered molybdenum–vanadium steel. For comparison, the curves for the normalized steel are included, as are also the curves for 0·5% molybdenum steel and carbon steel.

It will be noted that for carbon steel the difference in temperature between 0·1 and 0·2% of deformation in 100,000 hr. is negligible except at temperatures below about 450° C. For 0·5%

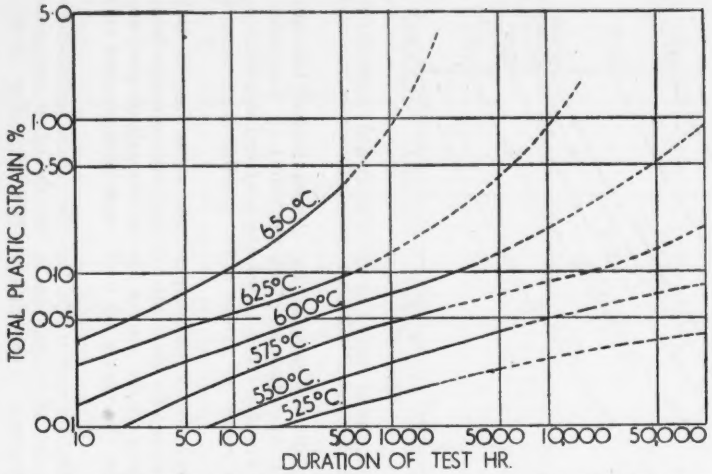


Fig. 65—Log.-strain/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at various temperatures indicated. Stress, 4 tons/sq. in.

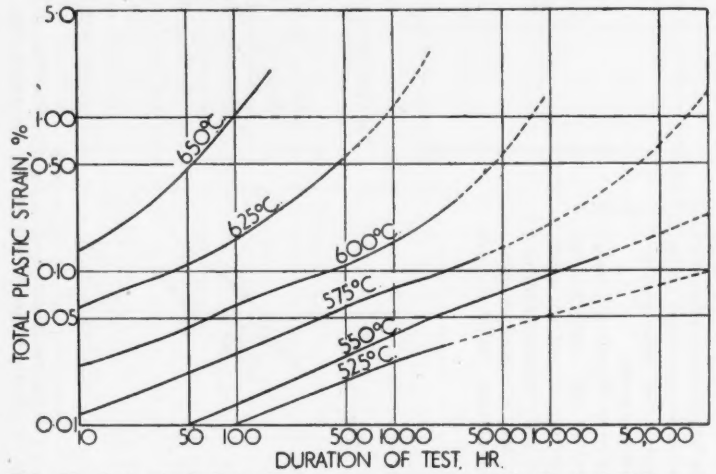


Fig. 66—Log.-strain/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at various temperatures indicated. Stress, 6 tons/sq. in.

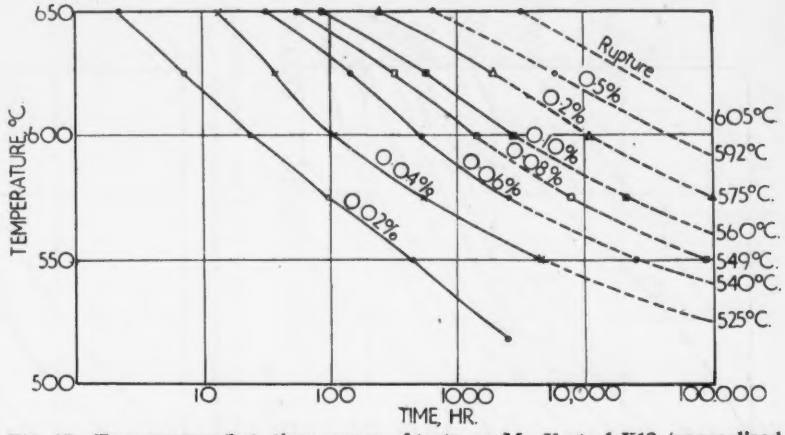


Fig. 67—Temperature/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at plastic strains as indicated. Stress, 4 tons/sq. in.

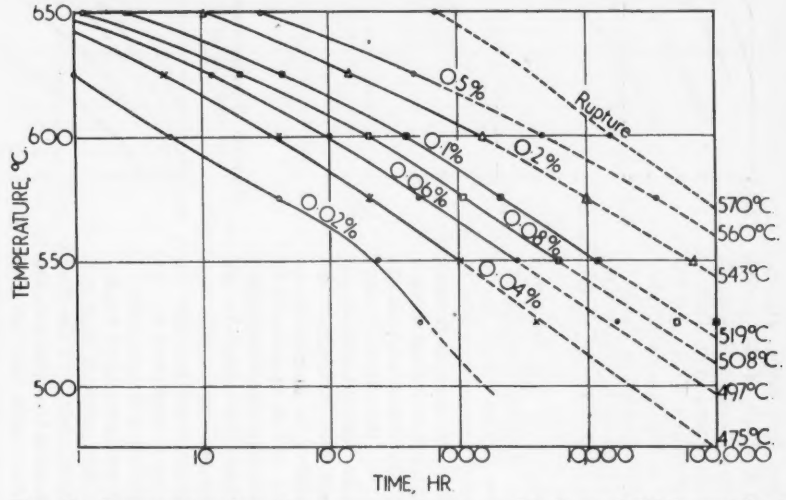


Fig. 68—Temperature/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at plastic strains indicated. Stress, 6 tons/sq. in.

molybdenum steel the difference is noticeable but becomes very small at the higher temperatures. With molybdenum–vanadium steel, on the other hand, there is a marked difference in temperature between 0·1 and 0·2% of deformation, which shows that if 0·1% of deformation in 100,000 hr. is the criterion of working stress, there is a greater factor of safety with this steel than with the others.

DISCUSSION

The stress/temperature relationship for 0·1 and 0·2% of creep for molybdenum–vanadium steel in the normalized condition and in the normalized and tempered condition has been obtained and is compared with similar data for 0·5% molybdenum steel and carbon steel. Molybdenum–vanadium steel, particularly in the tempered condition, is much superior to the other types of steel.

It has also been demonstrated that the long-time creep tests used to obtain the above type of relationship must be of sufficient duration to define the shape of the temperature/time curves of iso-strain, otherwise there is a possibility of the creep properties being underestimated.

A very important conclusion was reached regarding the effect of tempering on the creep properties of molybdenum–vanadium steel, namely, that tempering lowers the creep strength at high temperatures but it vastly improves the creep strength at lower temperatures, *i.e.*, within the range of temperature which is of practical importance. This effect is probably of minor importance in relation to carbon steels or 0·5% molybdenum steel, but for more complex steels it is of paramount importance. For example, as was shown by the author in the discussion of Oliver and Harris' paper[16] on austenitic steels for the gas turbine, part of their data, replotted in a different way to show the temperature/time relationship for 0·2% of creep at various stresses, indicates that a complex austenitic

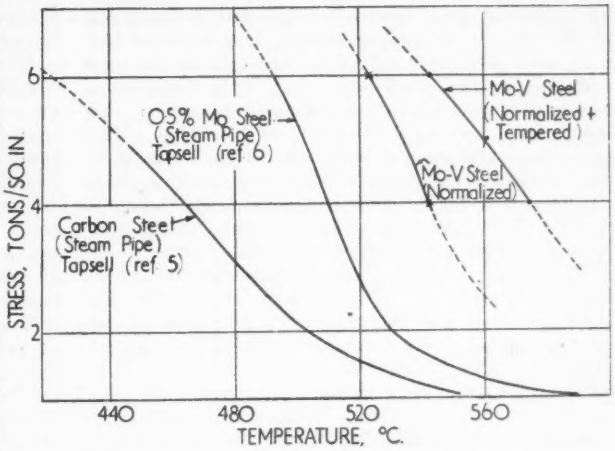


Fig. 69 (a)—Stress/temperature curves of tests on various steels for 0·2% of plastic strain in 100,000 hr.

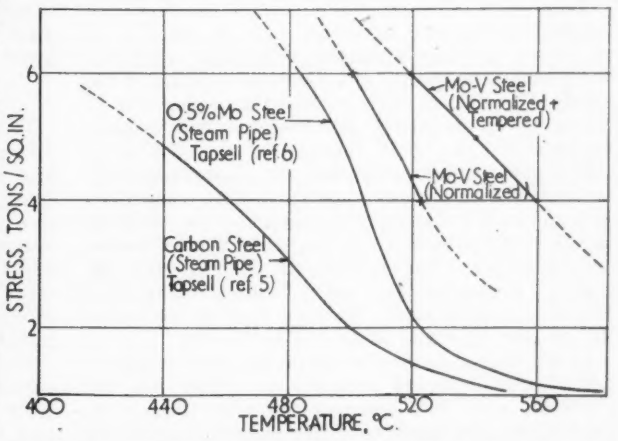


Fig. 69 (b)—Stress/temperature curves of tests on various steels for 0·1% of plastic strain in 100,000 hr.

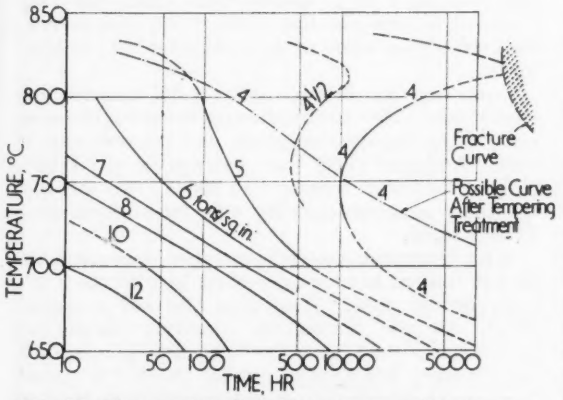


Fig. 70—Temperature/log.-time curves of tests on a complex austenitic steel (*see* ref. 16) for 0·2% of plastic strain at stresses indicated

steel actually has better creep properties at high temperature than at lower temperature, *i.e.*, the reversed **S** shape of the iso-stress curves is particularly pronounced, as shown in Fig. 70. If their complex steel can be 'tempered' to flatten out the reversed **S** shape, then the properties of the steel at lower temperatures and longer times will be greatly enhanced in a way suggested in Fig. 70. At the same time the ductility should improve considerably.

In Fig. 70 it will be noted that at a stress of 4 tons/sq. in. the 'tempered' steel is much inferior to the standard steel when tested at 825° C., but that at temperatures below about 750° C. the tempered steel is much superior. This illustrates an important point, namely, that the use of unsuitable short-time tests at high temperature and stress to evaluate the order of steels for long-time creep properties may, in the extreme case, indicate as the worst steel, that which actually would be the best in practice.

One further point to be emphasized is that, since

Fig. 66—Log.-strain/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at various temperatures indicated. Stress, 6 tons/sq. in.

Fig. 68—Temperature/log.-time curves of tests on Mo–V steel K12 (normalized at 975° C. plus tempering for 5 hr. at 690° C., A.C.) at plastic strains indicated. Stress, 6 tons/sq. in.

stress/temperature curves for specified deformation in 100,000 hr. require exceedingly long times of test and, since such tests only represent one cast of steel and only in the condition tested, it is advisable to ensure that such tests are carried out only on suitable steels in a properly heat-treated condition. This point naturally was not fully appreciated in the past, but in view of the possible range of properties which are obtainable for steels which are nominally the same, and in view of the importance of heat-treatment, it should be more fully considered in the future.

Considering the stress/temperature data shown in Figs. 2 (*a* and *b*) it is clear from the present and previous work that the poor creep properties of one of the carbon steels (superheater tube) was due to the excess addition of aluminium to the steel. Two of the 0·5% molybdenum steels also show poor creep properties. One of them (the cast steel) was furnace cooled from the normalizing temperature and was then given a prolonged tempering treatment. The other 0·5% molybdenum steel (the superheater tube) was open annealed from 700° C. Both of these treatments explain the poor creep properties obtained.

Once the working-stress/temperature data has been established for any particular type of steel, it is necessary to adopt a suitable short-time creep test for proving the quality of subsequent supplies of the steel. For carbon steels a short-time creep test of 8 tons/sq. in. and 450° C. has been specified for plate material.[17] The main function of this test is to detect abnormality, and for this purpose it is eminently suitable for carbon steels, provided that comparison is made with a standard steel and that the extent of abnormality possible in any particular type of carbon steel is known. For example, the limiting creep rate for high manganese steel would be much less than for a low manganese steel. This has been demonstrated by the author,[11] and illustrated in Fig. 26.

A short-time creep test for proving the quality of alloy steels must of necessity be rather different because, as has been demonstrated, a short-time creep test in the normalized condition may actually favour the steel of lowest creep resistance at the higher working temperatures. The only reliable test, therefore, is one which will show up any possible deterioration. For 0·5% molybdenum steel or chromium–molybdenum steel a suitable test would be a creep test at 6 tons/sq. in. and 550° C., after a previous treatment of, say, 50 hr. at 650° C. A permissible deformation could be set which must not be exceeded in, say, 100 hr. This test simulates to some extent the effect of time at the service temperature. If the results are unsatisfactory it is probable that in service some part of the equipment will be inferior due to variations in heat-treatment and tempering effects caused by welding, etc. It must be remembered also that a very short time at temperatures above 700° C. will be equivalent to quite a long time at 650° C.

For molybdenum–vanadium steel or more complex steels which are used in the tempered condition, the short-time creep test must be one which makes a comparison between an actual creep curve at some particular temperature and stress, and a standard curve. For example, for molybdenum–vanadium steel a creep test at a stress of 4 tons/sq. in. and a temperature of 600° C. for a period of 100 or more hours is suitable.

However, before leaving this question of quality tests it must be emphasized that it is not necessary to be unduly stringent. It is very difficult, if not impossible, to obtain exactly similar short-time creep tests from cast to cast of a particular type of steel. The differences obtained affect particularly the variation in the amount of creep in the intial stages of the test, since apparently this is influenced by quite minor variations of composition and heat-treatment. If, when the initial stage of creep is past, the creep curve of the steel runs approximately parallel to the standard curve, then the steel is of good quality. If, however, the steel continues to deviate markedly from the standard curve then something has to be corrected. It must be realized that the variations obtained in a short-time test which is carried out at a high temperature and stress are an exaggeration of what will be obtained in service. At the most, it may be said that such variations will result in the steel deforming 0·11 or possibly 0·12% instead of 0·10% in 100,000 hr. of actual service. A considerable number of tests have been carried out to explain the above point which will be dealt with more fully in a later paper.

MECHANICAL PROPERTIES

There is no direct relationship between the mechanical properties of steel as ordinarily measured either at high temperatures or at room temperature, and the creep properties. However, since the room-temperature properties are of considerable importance for other reasons, these have been included in Table IX, which shows the mechanical properties of the three main types of steel, namely, 0·5% molybdenum steel, 0·5%-molybdenum/0·8%-chromium steel, and 0·5%-molybdenum/0·25–0·3%-vanadium steel.

With 0·5% molybdenum steel, increasing the carbon content increases the ultimate stress and yield point and decreases the elongation, reduction of area, and impact strength. Increasing the molybdenum content from 0·5 to 1·33% increases the ultimate stress slightly, and slightly decreases the yield point. As is indicated in Table IX, the limit of proportionality of the 1·33% molybdenum steel was only 6 tons/sq. in., whereas that of the 0·5% molybdenum steel was almost equal to the yield point, *i.e.*, 18 tons/sq. in.

Tempering the 0·09%-carbon/0·5%-molybdenum steel at 650° C. lowers the ultimate stress but increases the yield point, the elongation, and the reduction of area. A similar effect was observed for the 1·33% molybdenum steel tempered at 690° C. In this case the limit of proportionality was raised from 6 to 21 tons/sq. in.

With chromium–molybdenum steel, increasing the carbon content has the same effect as with the 0·5% molybdenum steel. Tempering also has a similar effect. In the normalized condition chromium–molybdenum steel has a low limit of proportionality, for example, 8·6 tons/sq. in. for the 0·1% carbon steel, which rises on tempering to equal the yield point. As shown above, a similar effect was obtained with the 1·33% molybdenum steel.

Molybdenum–vanadium steel behaves like the others with increasing carbon content when normalized at 950° C. or above. The limit of proportionality was high and almost equal to the yield point, but when normalized at 850° or 900° C. the limit of proportionality was only 8·0 or 11·7 tons/sq. in. respectively. The impact strength of molybdenum–vanadium steel decreases with increasing normalizing temperature and is very low when the normalizing temperature is above 1000° C., due to the increasing coarseness of the structure.

On tempering the normalized molybdenum–vanadium steel for a short time at 690° C. the ultimate stress is raised as also is the yield point and reduction of area. This is due to precipitation hardening effects. Further tempering gives a gradual fall in the ultimate stress and yield point and a rise in the reduction of area. The impact strength improves with tempering although there may be a drop if the tempering treatment is too short. This is more noticeable the higher the normalizing temperature.

It is clear from the above that the effect of heat-treatment on the mechanical properties of the above types of steel is by no means simple. Testing at high

Table IX

MECHANICAL PROPERTIES OF MOLYBDENUM, CHROMIUM–MOLYBDENUM, AND MOLYBDENUM–VANADIUM STEELS

Steel	Alloying Elements			Heat-Treatment	Mechanical Properties					
					Ultimate Stress, tons/sq. in.	Elastic Limit, tons/sq. in.	Yield Point, tons/sq. in.	Elongation (on $4\sqrt{\text{Area}}$), %	Reduction of Area, %	Izod Impact, ft. lb.
				Molybdenum Steels						
	C, %		Mo, %							
A2	0·09		0·51	Normalized at 950° C.	29·3	17·6	17·9	35·0	68·0	80, 86, 82
				Normalized + 10 hr. at 650° C., A.C.	27·6	18·6	18·6	37·0	72·0	..
				,, + 100 ,, ,, ,,	25·9	18·0	18·0	42·0	79·0	..
A3	0·145		0·48	Normalized at 950° C.	31·8	..	16·5	32·0	58·0	65, 68, 66
A4	0·21		0·55	,, ,, 920° C.	39·2	..	29·0	30·0	56·0	36, 39, 38
A5	0·31		0·52	,, ,, 920° C.	44·6	..	31·2	22·0	48·0	16, 15, 12
A8	0·12		1·33	,, ,, 900° C.	32·7	8·4	17·4	29·0	63·0	56, 60, 58
				,, ,, 950° C.	33·8	6·0	16·2	28·0	58·0	45, 42, 40
				Normalized + 5 hr. at 690° C.	30·4	21·0	21·3	35·0	73·0	..
A9	0·13		0·55	Normalized at 950° C.	29·4	17·8	18·0	36·0	64·0	87, 76, 96
				Chromium–Molybdenum Steels						
	C, %	Mo, %	Cr, %							
B1	0·10	0·53	0·75	Normalized at 950° C.	28·7	8·6	16·0	34·0	71·0	82, 84, 86
				Normalized + 10 hr. at 650° C.	27·1	19·3	19·3	37·0	75·0	..
				,, + 50 ,, ,, ,,	27·3	19·5	19·5	36·0	75·0	..
				,, + 100 ,, ,, ,,	26·2	19·0	19·0	40·0	76·0	..
B2	0·145	0·54	0·75	Normalized at 950° C.	35·0	11·4	21·2	29·0	59·0	58, 60, 61
B3	0·22	0·57	0·48	,, ,, 880° C.	43·0	..	29·4	25·0	53·0	32, 28, 27
B4	0·125	0·515	0·88	,, ,, 950° C.	35·2	9·4	20·9	25·0	64·0	77, 73, 74
B6	0·12	0·585	0·82	,, ,, 950° C.	32·8	9·8	19·7	31·0	59·0	85, 83, 86
				Molybdenum–Vanadium Steels						
	C, %	Mo, %	V, %							
K 2	0·08	0·29	0·11	Normalized at 1000° C.	27·7	19·0	19·2	34·0	68·0	95, 97, 95
K 3	0·085	0·29	0·21	,, ,, 1000° C.	28·1	18·8	19·6	34·0	69·0	86, 80, 82
K 4	0·08	0·30	0·32	,, ,, 1000° C.	27·9	19·3	19·3	34·0	69·0	101, 93, 97
K 6	0·105	0·60	0·10	,, ,, 950° C.	29·6	20·0	20·8	38·0	73·0	89, 86, 94
K 7	0·115	0·595	0·215	,, ,, 950° C.	30·4	20·5	21·5	33·0	72·0	73, 73, 82
K 8	0·12	0·585	0·31	,, ,, 950° C.	31·5	21·5	22·4	29·0	68·0	74, 72, 72
K 9	0·19	0·53	0·31	,, ,, 950° C.	44·4	26·0	29·2	24·0	54·0	4, 5, 5
,,	,,	,,	,,	Normalized + 1 hr. at 690° C., A.C.	40·4	29·4	29·4	26·0	64·0	20, 23, 25
K10	0·13	0·62	0·31	Normalized at 950° C.	35·8	21·5	21·8	31·0	67·0	52, 60, 67
,,	,,	,,	,,	Normalized + 4 hr. at 690° C., A.C.	36·5	24·3	24·3	29·0	74·0	119, 119, 118
,,	,,	,,	,,	,, + 20 ,, ,, ,, ,,	35·3	23·2	23·2	30·0	75·0	117, 118, 119
,,	,,	,,	,,	,, + 40 ,, ,, ,, ,,	34·8	23·8	23·8	30·0	77·0	111, 115, 113
K12	0·12	0·5	0·255	Normalized at 850° C.	31·4	8·0	16·0	36·0	75·0	47, 45, 37
,,	,,	,,	,,	,, ,, 900° C.	32·4	11·7	18·4	33·0	75·0	21, 18, 25
,,	,,	,,	,,	,, ,, 950° C.	35·6	23·5	24·1	30·0	71·0	27, 25, 25
,,	,,	,,	,,	,, ,, 1000° C.	37·6	24·0	25·5	21·0	67·0	15, 18, 14
,,	,,	,,	,,	,, ,, 1050° C.	39·2	23·5	25·8	22·0	62·0	3, 3, 3
,,	,,	,,	,,	,, ,, 950° C.	35·6	23·5	24·1	30·0	71·0	27, 25, 25
,,	,,	,,	,,	Normalized + 5 hr. at 690° C., A.C.	35·2	27·0	27·0	32·0	74·0	27, 29, 27
,,	,,	,,	,,	,, + 20 ,, ,, ,, ,,	34·8	26·8	26·8	33·0	76·0	35, 37, 39
,,	,,	,,	,,	,, + 50 ,, ,, ,, ,,	33·7	25·2	25·2	33·0	77·0	44, 42, 45
,,	,,	,,	,,	Normalized at 1000° C.	37·6	24·0	25·5	21·0	67·0	15, 18, 14
,,	,,	,,	,,	Normalized + 2 hr. at 690° C., A.C.	40·4	33·6	33·6	25·0	72·0	9, 8, 7
,,	,,	,,	,,	,, + 10 ,, ,, ,, ,,	39·2	31·2	31·2	27·0	74·0	22, 26, 26
,,	,,	,,	,,	,, + 50 ,, ,, ,, ,,	34·8	27·6	27·6	28·0	76·0	34, 32, 38

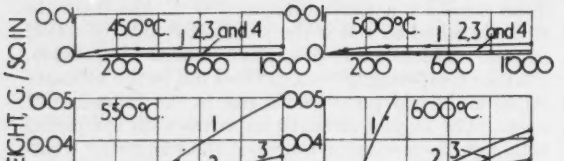


Fig. 71—Results of scaling tests in air at 450°, 500°, 550°, and 600° C., on mild steel and steels A2, B1, and K12 (curves 1, 2, 3, and 4, respectively)

temperature further complicates the problem. Further consideration of this aspect of the problem will be dealt with elsewhere.

EFFECT OF SERVICE TEMPERATURE

EMBRITTLEMENT

Bailey and Roberts[3] were among the first to show that serious embrittlement, as measured by an impact test at room temperature, may result when a steel is heated for long periods at temperatures of about 400° to 600° C. although, apparently, the mechanical properties other than impact are not affected. Since then the subject has been explored extensively, and it appears that this type of embrittlement is confined to certain types of steel usually having high nickel, manganese, or copper contents.

Chromium–molybdenum steel and 0·5% molybdenum steel both in the normalized condition, and molybdenum–vanadium steel in the normalized and tempered condition, have been tested after heating for periods up to 3000 hr. at 400°–600° C., and there is no suggestion of any embrittlement effect. It is therefore unnecessary to tabulate any of the results obtained. As already shown, in the normalized condition molybdenum–vanadium steel may show some embrittlement by tempering for a short time at 690° C. At lower temperatures a long-time tempering treatment will have the same effect, so that it may be said that normalized molybdenum–vanadium steel will be subject to embrittlement due to the effect of time at the service temperature. This effect, however, which is due to precipitation hardening is accompanied by an increase in the ultimate stress and yield point. However, by suitable tempering at 690° C. this effect is easily overcome.

SCALING

Very little information is available in the literature on the effect of prolonged heating and the degree of scaling obtained on tests carried out in air or a steam atmosphere. At temperatures below 600° C. scaling is not so severe as has been generally believed. The reason for this is probably because references to scaling have been made in connection with rupture tests which have failed in an intercrystalline manner. Since the cracks penetrating from the surfaces are filled with oxide, the impression has been given that the scaling is severe.

Figure 71 shows results of scaling tests in air, carried out for periods of up to 1000 hr. exposure at 450°, 500°, 550°, and 600° C. It will be noted that at 450° and 500° C. scaling is negligible, although mild steel shows a greater attack than does 0·5% molybdenum steel, chromium–molybdenum steel, or molybdenum–vanadium steel.

At 550° C., and particularly at 600° C., the attack on the mild steel is proceeding at quite a high rate after 1000 hr. exposure, whereas with the other steels the rate has slowed down considerably. At 550° and 600° C. molybdenum–vanadium steel appears to be slightly better than 0·5% molybdenum steel or chromium–molybdenum steel.

In Fig. 72 an approximate curve for molybdenum–vanadium steel is shown for the penetration loss from a surface in air at 550° C. for up to 50,000 hr. This is based on measurements taken at the shoulders of the rupture test-pieces at 550° C. The 50,000 hr. result was taken from the test at 4 tons/sq. in., and due to the negligible amount of creep the same result was obtained on the gauge length of this test-piece.

Rohrig and others[18] have carried out tests in steam on many types of steel for periods up to 16,000 hr. at 593° C. (1100° F.). The results are reproduced in Fig. 72. It was found that the most severe attack occurred on carbon steel. Low-alloy steel, varying from 0·5% molybdenum steel to 5·0% chromium steel containing 0·5% of molybdenum and 1·5% of silicon, showed a much greater resistance to corrosion. It was shown also, that the higher chromium steels (up to 5% of chromium) were no better than 0·5% molybdenum steel. In view of the results in air discussed above, it can be inferred that molybdenum–vanadium steel will also fall in this group. Stainless steel showed much greater resistance to corrosion by steam at this temperature.

It may be said, therefore, that scaling in steam is no more severe than is scaling in air, and that at 550° C. the rate of attack on low-alloy steels is not serious.

Agnew and others[19] carried out rupture tests for periods of up to 8000 hr. on 9 different types of steel varying from carbon steel to 18/8 stainless steel in steam at 1200° F., and found that the rupture strength was greater than in air at the same temperature and stress.

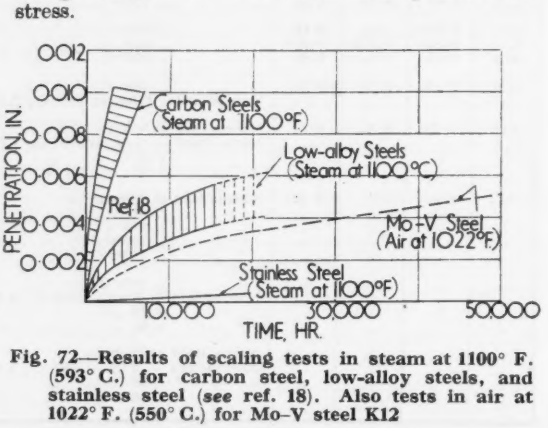


Fig. 72—Results of scaling tests in steam at 1100° F. (593° C.) for carbon steel, low-alloy steels, and stainless steel (see ref. 18). Also tests in air at 1022° F. (550° C.) for Mo–V steel K12

GENERAL DISCUSSION AND CONCLUSIONS

The many variables which exert an appreciable influence on the creep of steel under stress at temperatures above 700° F. make the study of creep exceedingly complex, and for the same reason it is impracticable to repeat at length all the conclusions embodied in this paper. Comparatively minor differences in composition, degree of deoxidation, or heat-treatment acquire an importance not hitherto fully appreciated, whilst the effect of varying the test conditions must be studied in relation to the particular steel variable which is being considered. To determine the stress/temperature relationship for a specified deformation in 100,000 hr., it is therefore necessary to carry out an extensive series of long-time creep tests, and the results obtained can be applied strictly only to the particular sample in the particular condition tested. It is important, therefore, to ensure that such tests are carried out only on suitable steels in a properly heat-treated condition. Thus some form of short-time test is necessary, first as a preliminary examination of the effects of composition and treatment variables on a given type of steel and secondly, as a check on the quality of steel of the same type as the prototype for which the stress/temperature relationship has been determined.

With regard to the effects of composition and treatment variables, the more complex the steel and the less the information available for similar types of steel, the more testing is required to determine the best treatment. Steels which have excellent creep properties are almost certain to be benefited by tempering, particularly when they are required for very long time service conditions. It is also advisable to carry out preliminary rupture tests, but due cognizance must be taken of the fact that these are generally carried out at a higher temperature and/or stress than will be the conditions in service.

As stated in the introduction, Bailey and Roberts[3] formulated a law relating to carbide spheroidization, as follows:

$$t = Ae^{b/T},$$

where t = time, T = absolute temperature, and A and b are constants; *i.e.*, the same structural change obtained by tempering for a given time at one temperature is obtained at any other temperature (below the critical point of the steel) in a time which can be calculated from the above formula.

Later, Bailey extended this work by relating the effect of previous tempering of creep test-pieces with the effect of time at the testing temperature on the long-time creep properties of the normalized steel. For example, as shown by the creep tests in Fig. 3, the material may be expected to be improved in creep resistance by the effect of temperature during service for 100,000 hr. at 475° C.

This classical idea was fully justified on the evidence then available, and the author gladly takes this opportunity of recording his debt to Bailey. His published work referred to above has been a source of inspiration and the author is surprised that so little use has been made of it in later work.

The above law will probably hold reasonably true for simple steels containing iron carbide only, but in more complex steels the carbides present in the steel in the normalized condition will alter their composition during tempering or due to the effect of time at service temperature. Bowman[20] has shown, with 0·5% molybdenum steel in the quenched and mildly tempered condition, that most of the molybdenum is in the ferrite and very little is in the carbide, whereas on tempering at higher temperatures the molybdenum reverts to the carbide. Thus, since 0·5% molybdenum steel or more complex steels in the normalized condition are not truly stable, it is to be expected that some change will occur in the carbide composition with tempering. This will invalidate the above relationship since the constant b will change with the composition of the carbide.

However, even assuming the above law to be exact, a further relationship connecting tempering time and creep properties is not so simple as would appear. It is true that if the creep tests carried out after various times of tempering at 650° C., shown in Fig. 3, had been repeated after tempering at say, 600° C. for equivalent times, the same curve would be obtained, but it cannot be assumed from this that tempering will have the same effect under different conditions of testing. For example, in Fig. 21 (*b*), showing creep tests at a stress of 6 tons/sq. in. and a temperature of 550° C. on chromium–molybdenum steel, the best creep resistance as measured by the time to reach 0·1% of deformation was obtained by tempering for 30 hr. at 650° C.; whereas at a stress of 4 tons/sq. in. and a temperature of 600° C., as shown in Fig. 22 (*b*), the tempering time for the best creep resistance was 20 hr. This cannot be accounted for by the additive effect of tempering due to the time at the testing temperature.

Again, for 0·5% molybdenum steel the best tempering time was 10 hr. for tests carried out at a temperature of 550° C. and a stress of 6 tons/sq. in. (Fig. 7 *b*); whereas in tests at a stress of 4 tons/sq. in. and a temperature of 600° C. (Fig. 12 *b*) the best result was obtained with no tempering treatment. Similarly, by carrying out creep tests at one temperature and at different stresses, it is found that the higher the stress the less the tempering time required to give the maximum creep resistance at 0·1% of deformation.

Thus, since the tempering time at which maximum stiffening occurs varies, depending on the particular test conditions used, the results obtained on one steel in any one kind of short-time test cannot be used to determine what will happen under long-time creep test conditions.

It must be emphasized, however, that Bailey's method of approach to the problem, as illustrated in Fig. 3, is an invaluable one in determining the potentialities of different steels. It also indicates that before carrying out short-time tests to demonstrate that a particular batch of steel is up to standard, the material should first be subjected to a suitable tempering treatment.

It has been proved that the effect of tempering any steel (within limits) is to lower the creep properties above a certain temperature (this temperature being lower the higher the stress and/or the greater the permissible deformation) and to improve the properties

below that temperature. It may happen, therefore, that for the service life intended (say 100,000 hr.) tempering is essential in order to develop the maximum creep resistance for a particular steel. As has been demonstrated, this applies to molybdenum–vanadium steel.

A further effect of tempering, which may make it essential even if it does not appreciably improve the creep properties, is that it improves the ductility of the steel at fracture and there is less chance of failure due to intercrystalline cracking.

A similar effect occurs by varying the normalizing temperature of the steel, *i.e.*, increasing the normalizing temperature increases the long-time creep properties at higher working temperatures and will decrease them at lower temperatures. For electric-furnace steels or high-frequency-furnace steels a higher normalizing temperature is necessary to give the same properties as an open-hearth steel and thus, in general, open-hearth steel is the best for steels in use at high temperatures.

The effect of varying the percentage of an element also influences the long-time creep properties in a way similar to that of tempering. Thus, for example, silicon is detrimental to the long-time creep properties of 0·5% molybdenum steel or chromium–molybdenum steels except at low working temperatures. Aluminium has a similar effect which, however, is more pronounced.

Vanadium, and to a lesser extent manganese and chromium (within limits), increase the long-time creep properties of molybdenum steel. Care must be taken in interpreting the effect of composition, since this effect is so dependent on the heat-treatment used.

These conclusions were not fully appreciated until much of the long-time tests on the molybdenum–vanadium steel had been carried out, although preliminary work had shown the advisability of a low carbon content and a tempering treatment for this type of steel. These long-time tests finally proved the principles by which short-time tests can be used to demonstrate whether one steel will be better than another under long-time service conditions, and also what form of short-time test is most suitable for checking the quality of a batch of steel of that particular type.

The stress/temperature relationship for 0·1 and 0·2% of deformation in 100,000 hr. for molybdenum–vanadium steel shows it to be superior to 0·5% molybdenum steel or chromium–molybdenum steel. The tempered molybdenum–vanadium steel is rather superior to the same steel in the normalized condition and shows greater ductility before fracture. The short-time creep tests confirm these results and demonstrate the fundamental importance of proper heat-treatment. Similarly, the stress/temperature relationship for rupture in 100,000 hr. also shows that molybdenum–vanadium steel is superior to 0·5% molybdenum steel or chromium–molybdenum steel.

The primary purpose of this paper has been to present essential information regarding the working-stress/temperature properties of molybdenum–vanadium type steel which has enhanced properties as compared with normal steels at present in use for high-temperature superheater and steam pipes, etc.

At the same time, since there was considerable confusion regarding the fundamental aspects of the subject, and since a great deal of published data was liable to be misrepresented, it was considered essential to attempt to clarify this position so that the various effects could be seen in their proper perspective.

No attempt has been made in this paper to discuss the fundamental reasons for the different behaviour of the various steels under creep testing conditions. On the basis of creep tests alone, it would be almost impossible to demonstrate these fundamental effects which are included in the more general problem of the deformation of metals at any temperature. A great deal of further work has been carried out on the above and other steels with reference to this more general problem, and it is hoped to present this work in the near future.

Acknowledgments

The author gratefully acknowledges the valuable assistance of his colleagues, particularly Mr. A. J. K. Honeyman, Deputy Chief Metallurgist, and he also desires to express his thanks to Mr. W. Barr, Executive Director and Chief Metallurgist, Messrs. Colvilles, Ltd., for his continued encouragement and advice, and to him and his fellow Directors for permission to publish this paper.

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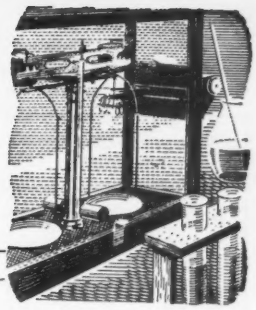
BRITISH IRON AND STEEL RESEARCH ASSOCIATION



Mould-Weight/Ingot-Weight Ratio and its Relation to Mould Consumption

By N. H. Bacon, A.Met., F.I.M.

SYNOPSIS

An examination has been made of the very comprehensive data provided by three Ingot Moulds Sub-Committee questionnaires (one pre-war and two post-war), and British Intelligence Objectives Sub-Committee Final Report No. 685, "German Ingot Moulds for the Casting of Steel Ingots."

The data examined were for many types and sizes of moulds used in the various works in Great Britain and Germany, and from them the importance of the mould-weight/ingot-weight ratio in its relation to mould consumption has been established. Attention is drawn to the very large variations in mould-weight/ingot-weight ratio found within the United Kingdom and Germany, *viz.*, in the United Kingdom 0·78 to 1·76 and in Germany 0·61 to 1·82, and possible reasons for such variations are suggested. A graph is given showing what appears from the data to be the optimum ratio for moulds of capacities of more than 18 cwt.

Introduction

THE performance and life of moulds for steel ingots depend on the conditions of manufacture and use of the moulds, and on their design. This paper deals only with one aspect of mould design, the ratio of the weight of the mould to that of the ingot for which it is used (hereafter referred to as the M/I ratio). It sets out to examine the relationship between this ratio and the consumption of mould material per ton of steel produced, this being economically a more important consideration than mould life. This matter has not been dealt with at all exhaustively in previous papers. Ristow[1] gives curves obtained from data collected in Germany showing maximum mould consumption with M/I ratios around 1·25. The effects he shows are not large, and are commented on only briefly. Reagan[2] describes tests on a series of moulds of different weights, all for one type of ingot, in which the M/I ratio was increased from 1·79 to 2·34, the mould consumption decreasing slightly. Several references have also been made to the effect of mould wall thickness on mould life, *e.g.*, Swinden and

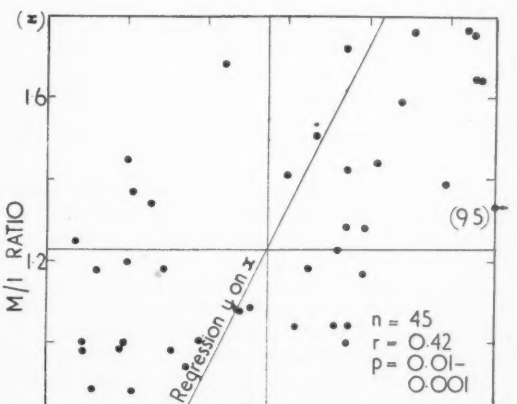


Fig. 1—Correlation between mould consumption and M/I ratio for British moulds of all types. (Data from the First I.M.S.C. Report.) Regression : 0·1 increase in M/I ratio increases mould consumption by 2·56 lb. per ton of steel

Paper SM/BA/50/47 of the Ingot Moulds Sub-Committee of the Steelmaking Division of the British Iron and Steel Research Association. The views expressed are the author's and are not necessarily endorsed by the Sub-Committee as a body.

The manuscript was received on 13th September, 1947.

Mr. Bacon is employed by the Steel, Peech, and Tozer Branch of the United Steel Companies, Ltd., Sheffield.

Table I

DATA FROM FIRST REPORT OF THE I.M.S.C.

Mould No.	Mould Weight, cwt.	M/I Ratio	Mould Consumption, lb./ton
A	63	1·00	24·7
B 1	26	1·76	72·0
B 2	32	1·64	73·7
B 3	26	1·76	72·0
B 4	32	1·64	73·7
B 5	72	1·76	65·4
B 6	174	1·08	45·2
B 7	280	1·04	55·3
C 1	130	1·18	26·0
C 2	163	1·18	34·8
D	64	1·51	53·7
E 1	48	1·04	51·0
E 2	66	0·98	35·3
E 3	96	1·17	59·2
E 4	88	1·01	39·0
E 5	170	1·00	57·0
F	53	1·39	69·5
G 1	320	1·23	56·0
G 2	450	1·18	52·6
G 3	720	1·14	57·2
H 1	58	1·72	57·0
H 2	98	1·41	49·8
J 1	60	1·34	33·0
J 2	123	0·84	45·0
K 1	35	1·28	57·0
K 2	35	1·28	59·5
K 3	293	0·88	30·2
K 4	1031	0·84	35·5
K 5	2120	0·94	37·2
L 1	21	1·33	95·0
L 2	38	1·44	61·0
M 1 (i)	164	0·82	42·5
M 1 (ii)	178	0·89	25·7
M 2 (i)	80	1·00	29·5
M 2 (ii)	80	1·00	29·5
N	64	0·98	24·5
O	57	1·42	57·0
P	30	1·68	42·0
Q 1	22	1·59	64·0
R	60	1·20	30·0
S	50	1·25	23·5
T 1	36	1·45	30·0
T 2	145	1·08	44·0
U	50	0·78	57·0
W	88	1·37	30·5
Mean		1·23	47·6

Bolsover,[3] and the First Report of the Ingot Moulds Sub-Committee.[4] The literature does not, however, provide any generally applicable recommendations on M/I ratio.

Much data on ingot moulds has been collected at various times in this country by means of questionnaires sponsored by the Ingot Moulds Sub-Committee (hereafter referred to as the I.M.S.C.), and data is also available for a large number of moulds used in Germany.

DATA FROM THE FIRST REPORT OF THE INGOT MOULDS SUB-COMMITTEE

The relevant data provided in the First Report of the I.M.S.C.[5] are set out in Table I, and mould consumption is plotted against M/I ratio in Fig. 1. As was to be expected there is a very wide scatter of

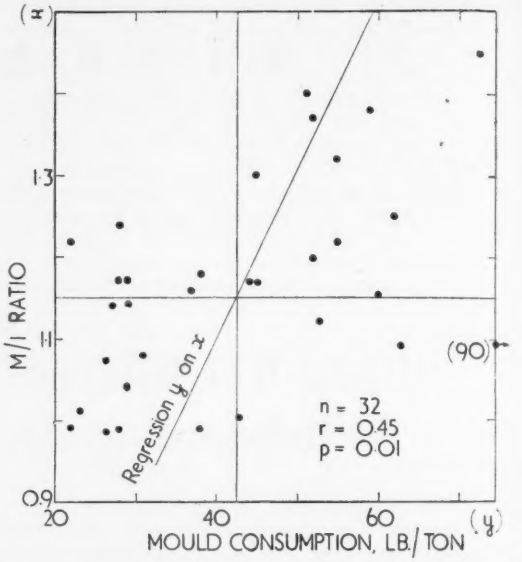


Fig. 2—Correlation between M/I ratio and mould consumption for British moulds of all types. (Data from I.M.S.C. questionnaire, 1945). Regression: 0·1 increase in M/I ratio increases mould consumption by 5·45 lb./ton

the points because of the numerous operative variables other than the M/I ratio. The strength of the correlation of M/I ratio with mould consumption is, however, very marked—the odds against the result being fortuitous are between 100 and 1000 to 1.

The regression of the mould consumption on the M/I ratio is shown in Fig. 1, and amounts to 2·56 lb. per ton of steel for each 0·1 increase in M/I ratio. It will be noted that the data include a big variety of mould types and weights, *viz.* :

Small- and big-end-up moulds.
Moulds with and without feeder heads.
Squares, slabs, and octagons.
Moulds 21–2120 cwt. in weight.

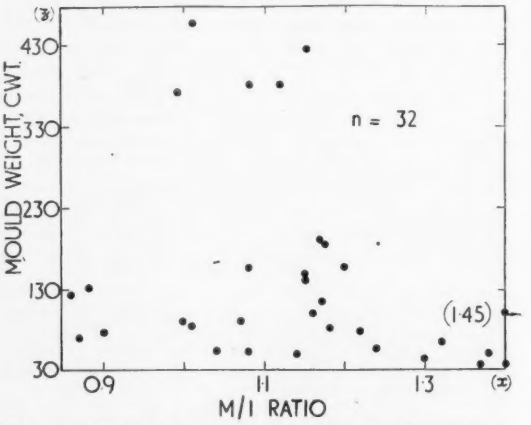


Fig. 3—Mould weight plotted against M/I ratio for British moulds of all types. (Data from the I.M.S.C. questionnaire, 1945)

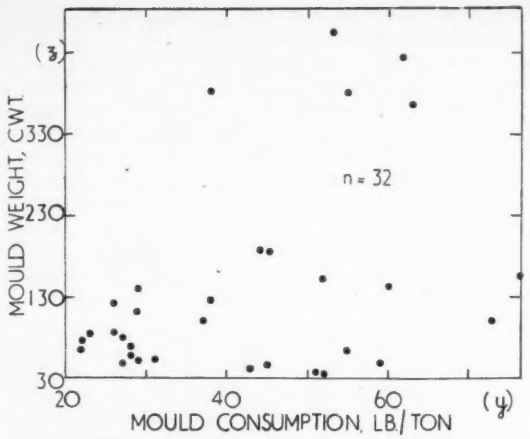


Fig. 4—Mould weight plotted against mould consumption for British moulds of all types. (Data from I.M.S.C. questionnaire, 1945)

It is also important to note that the mould consumption is, in each case, the average consumption over a long period.

DATA FROM I.M.S.C. QUESTIONNAIRE NO. 1 IN 1945

In Table II the data shown have been taken from the replies to the questionnaire issued by the I.M.S.C. in 1945. These data cover very similar ground to that covered by the I.M.S.C. First Report (*see* above), but there is an interval of about nine years between the two.

The correlation between M/I ratio and mould consumption is again very marked (*see* Fig. 2), and the regression of consumption on M/I ratio is, as will

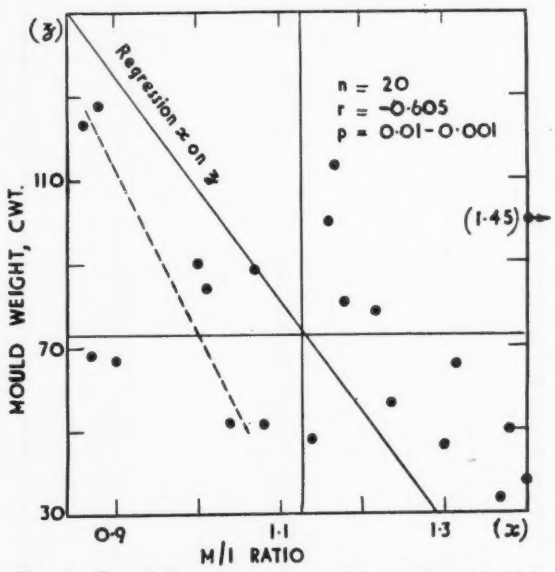


Fig. 5—Correlation between mould weight and M/I ratio for moulds of less than 130 cwt.

Table II

DATA FROM I.M.S.C. QUESTIONNAIRE NO. 1, 1945

Mould Weight, cwt.	M/I Ratio	Mould Consumption, lb./ton
Moulds of Less than 130 cwt.		
128	0·88	38
68	0·87	28
123	0·86	26
67	0·90	22
90	1·00	43
52	1·08	31
52	1·04	29
84	1·01	23
89	1·07	26
100	1·16	37
80	1·18	27
49	1·14	27
113	1·17	29
57	1·24	28
79	1·22	22
66	1·32	55
33	1·37	52
46	1·30	45
50	1·38	59
38	1·40	51
(100)	(1·45)	(73)
Mean* 73·2	1·13	34·9
Moulds of More than 130 cwt.		
368	1·09	63
155	1·08	90
181	1·17	45
147	1·15	60
458	1·12	53
187	1·17	44
380	1·18	38
140	1·15	29
380	1·22	55
153	1·20	52
425	1·25	62
Mean for all moulds	1·15	42·6

* Excluding the abnormal values in brackets

be seen, 5·45 lb. per ton of steel for each 0·1 increase in the M/I ratio. Comparing the two periods, pre- and post-war, it will be noted that the post-war M/I ratio is 0·08 lower (1·15 post-war against 1·23 pre-war), and also that the mould consumption is 5 lb./ton lower (47·6 pre-war, 42·6 post-war).

When this close relationship between M/I ratio and mould consumption had been established, it seemed desirable to try to determine the cause—were these two variables, M/I ratio and consumption dependent upon a third variable, *e.g.*, mould weight ?

The data from the 1945 questionnaire were therefore plotted as mould weight against M/I ratio, Fig. 3,* and mould weight against mould consumption, Fig. 4. It will be seen from these two scatter diagrams that the moulds of more than 130 cwt. in weight are

* The spurious correlation to be expected in general from the use of a common factor, such as "mould weight," will be seen, by reference to Fig. 3, to be negligible here. This objection is in any event met by the partial analysis made and referred to later.

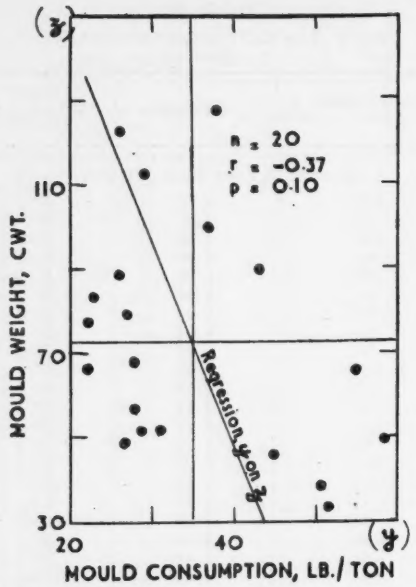


Fig. 6—Correlation between mould weight and mould consumption for moulds of less than 130 cwt.

quite out of line with those of less than 130 cwt. and so require separate examination. The data for the moulds weighing 130 cwt. or less were therefore plotted thus : Mould weight against M/I ratio ; mould weight against consumption ; and M/I ratio against consumption. *See* Figs. 5, 6, and 7. The correlation and regression results are shown and it will be seen that they are significant in Figs. 5 and 7, but there is one chance in ten that the regression in Fig. 6 is fortuitous.

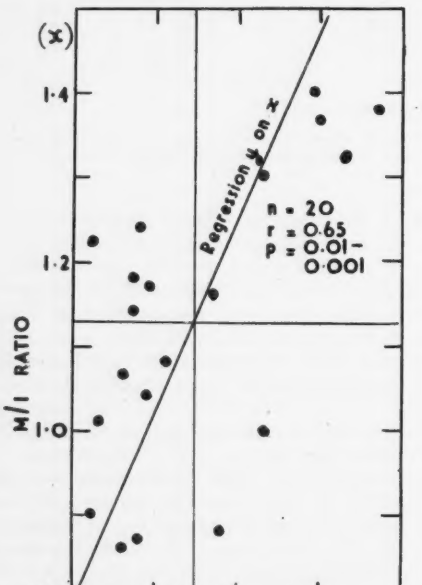


Fig. 7—Correlation between M/I ratio and mould consumption for moulds of less than 130 cwt.

By partial correlation it is possible to determine the effect of one variable while other variables are kept constant. In the case under discussion, the correlation between M/I ratio and mould consumption is still highly significant when the variation in mould weight is eliminated. On the other hand the correlation between mould weight and consumption becomes entirely insignificant when the effect of the variation in the M/I ratio is taken into account (*see* Table III, item 6).

It will be seen from Fig. 5, that the M/I ratio increases as the mould weight decreases and that nine of the samples have a lower M/I ratio and ten have

Table IV
DATA FOR PRE-WAR GERMAN MOULDS

Mould Weight, tons	M/I Ratio	Mould Consumption, lb./ton
2·9	1·11	26·1
3·5	1·20	17·2
3·5	1·20	24·6
3·5	1·13	19·0
3·4	1·10	19·2
4·9	1·41	17·2
4·1	1·13	13·7
4·8	1·30	21·1
4·6	1·21	27·0
5·0	1·21	17·1
5·5	1·19	20·2
5·7	1·18	20·4
4·5	1·25	27·1
4·8	1·31	22·2
5·2	1·36	31·0
5·0	1·31	28·6
4·2	1·07	13·6
5·3	1·26	27·2
5·5	1·12	22·4
5·5	1·10	28·6
6·3	1·21	20·4
6·0	1·13	28·6
1·0	1·33	23·7
1·1	1·44	22·0
1·7	1·05	24·4
2·5	1·11	27·5
2·8	1·15	27·5
3·3	1·27	22·7
2·6	1·01	22·0
3·5	1·10	31·0
3·8	1·18	17·2
4·8	1·29	25·2
4·1	1·10	16·0
5·1	1·35	19·8
4·4	1·16	26·8
4·6	1·21	29·8
4·2	1·00	22·5
5·0	1·17	16·8
5·0	1·16	26·2
5·5	1·18	28·5
4·3	1·23	34·0
4·8	1·33	25·8
4·5	1·23	32·0
4·9	1·32	29·7
4·7	1·25	36·0
4·8	1·26	25·4
4·2	1·07	16·0
5·5	1·10	32·4
5·8	1·16	34·4

a higher M/I ratio than that given by the regression line. The question arises, which of these two groups has the lowest mould consumption.

The nine moulds below the regression line average 0·98 M/I ratio and 29·7 lb./ton consumption, while the ten moulds above the line average 1·27 M/I ratio and 40·5 lb./ton consumption. It would therefore appear that the desirable M/I ratio lies to the left of the regression line shown in Fig. 5, and it is suggested that a regression line drawn from these nine moulds and shown as a broken line in Fig. 5 would be a better basis for deciding the desirable M/I ratio for any mould of between 50 and 130 cwt. capacity.

DATA FROM THE B.I.O.S. REPORT ON GERMAN INGOT MOULDS

Pre-War German Moulds

The data for pre-war German moulds set out in Table IV have been abstracted from B.I.O.S. Final Report No. 685, pp. 87–90. All the moulds given are included with the exception of six which were abnormal, *i.e.*, the first two on p. 87, the first two on p. 89, and the two hot-top moulds on p. 89. All except two of the moulds are of the small-end-up type. The data for M/I ratio and mould consumption are plotted in Fig. 8. The distribution of the points suggests a curvilinear relationship and this becomes more evident when the data are grouped in M/I ratio ranges thus:

M/I Ratio	Number of Types	Mould Consumption, lb./ton
1·00–1·09	5	19·70
1·10–1·14	11	24·05
1·15–1·19	9	24·22
1·20–1·24	8	25·26
1·25–1·29	6	27·27
1·30–1·34	6	25·18
1·35–1·45	4	22·50
	49	

These data are plotted in Fig. 9.

Table III

SUMMARY OF STATISTICAL DATA

Item	Source of Data	Number of Moulds	Class of Mould*	Range of Mould Weight, tons,	Mean Mould Weight, tons z	Mean M/I Ratio, x	Mean Mould Consumption, lb./ton, y	Correlation† r_{xy}	Level of Significance, p	Regression: Increase in Consumption on M/I Ratio for Each 0·1 Increase in M/I Ratio, lb.	Correlation‡ $r_{xy.z}$	Level of Significance, p	Correlation§ $r_{zy.x}$	Level of Significance, p
1	I.M.S.C. First Report	45	Various	1–106	9·0	1·23	47·6	0·42	0·01–0·001	2·56	...	...	...	...
2	I.M.S.C. Questionnaire, 1945, No. 1	32	,,	2–23	7·1	1·15	42·6	0·45	0·01	5·45	...	...	...	...
3	B.I.O.S. Report No. 685, pp. 88–90	49	All SEU, Open and SCT	1–6	4·3	1·20	24·2	0·12	Insignificant	...	...	...	...	...
4	As for 3; moulds of M/I ratio up to 1·25 only	35	,,	2–6	4·3	1·15	24·1	0·29	0·10	2·80	...	...	...	...
5	G.H.H. Moulds from B.I.O.S. Report No. 685, p. 80	16	Slab, SEU	1–16	4·1	1·31	42·9	0·74	0·001	3·40	0·90	0·001	0·01	No significance
6	Moulds of 6·5 tons or less from I.M.S.C. Questionnaire No. 1, 1945	20	Various	2½–6	3·6	1·13	34·9	0·65	0·01–0·001	4·60	0·57	0·02	< −0·01	,,
7	I.M.S.C. and B.I.O.S. Report No. 685	25	Octagons	4–160	34·4	0·90	39·6	0·01	Nil	...	0·003	Nil	0·04	,,
8	,, (excluding "G.H.H." moulds)	27	Slab, SEU	2–21	7·55	1·34	50·3	0·33	0·10	2·00	0·35	0·10–0·05	0·24	,,
9	Moulds of M/I ratio less than 1·34 included in item 8:					1·28	53·8							
	German	4	,,	8–10½	9·5	1·04	41·9							
	British	11	,,	4½–9	8·3									
	Moulds of M/I ratio more than 1·34 included in item 8:													
	German	12	,,	2–12	6·2	1·63	56·9							
	British	None												
10	Recent German moulds, B.I.O.S. Report No. 685	14	Squares, SEU	3–6	4·6	1·26	29·6	0·62	0·02	2·50	0·61	0·02	0·03	,,
11	I.M.S.C. First Report and Questionnaire No. 1, 1945	9	,,	2½–6	3·5	1·09	25·9	0·61	0·05	2·38	0·92	0·001	−0·85	0·01–0·001

* SEU = Small end up
SCT = Semi-closed top

† r_{xy} = Direct correlation between M/I ratio and mould consumption
‡ $r_{xy.z}$ = Correlation between M/I ratio and mould consumption with mould weight constant
§ $r_{zy.x}$ = Correlation between mould weight and mould consumption with M/I ratio constant

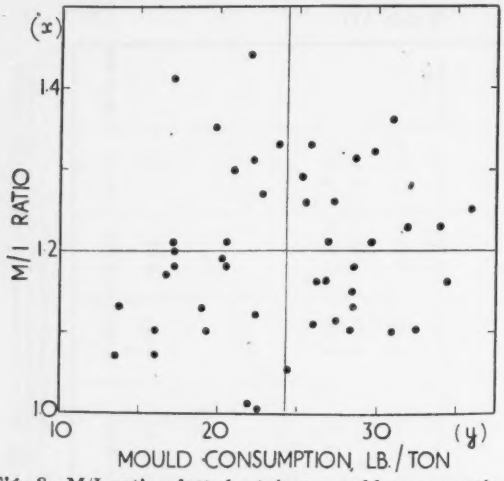


Fig. 8—M/I ratio plotted against mould consumption for pre-war German moulds

In view of this distribution, the correlation and regression were determined for all the moulds up to and including an M/I ratio of 1·25 (*see* item 4 of Table III, and Fig. 10). The regression shows 2·8 lb./ton increase in mould consumption for each 0·1 increase in M/I ratio for moulds up to 1·25 M/I ratio.

The apparent reduction in consumption for mould ratios over 1·25 is largely due to three items, *viz.*:

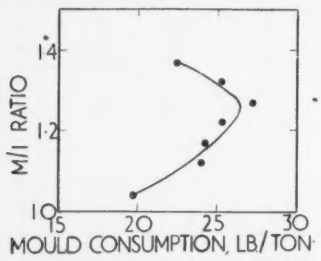


Fig. 9—Relationship between M/I ratio and mould consumption for pre-war German moulds

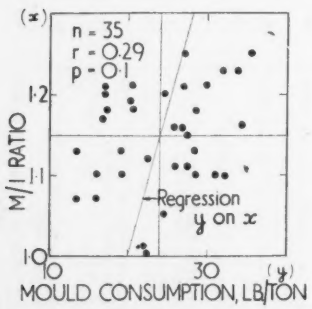


Fig. 10—Correlation between M/I ratio and mould consumption for German moulds of M/I ratio up to 1·25. Regression: 0·1 increase in M/I ratio increases mould consumption by 2·8 lb./ton

Table V

WARTIME GERMAN SQUARE SMALL-END-UP MOULDS

B.I.O.S. Report No. 685, Page	Mould Weight, tons	M/I Ratio	Mould Consumption, lb./ton	
76 and 25...	3·8	1·00	26·4	
77	4·0	1·33	32·5	
,, ...	4·5	1·18	21·8	
,, ...	4·8	1·26	23·3	
,, ...	5·2	1·30	31·8	
,, ...	4·3	1·07	19·7	
,, ...	5·8	1·09	26·8	
,, ...	5·3	1·00	18·5	
79 (Thomas steel) ...	3·1	1·41	35·9	*
,, ,, ,, ...	5·0	1·32	31·6	*
,, ,, ,, ...	5·0	1·39	40·6	*
79 (Open-hearth steel) ...	5·0	1·32	30·2	†
,, ,, ,, ...	5·0	1·39	37·4	†
82	4·1	1·64	38·3	
Mean	4·64	1·26	29·6	

* Average for years 1938 to 1943 inclusive
† Average for years 1938 to 1941 inclusive

	M/I Ratio	Mould Consumption, lb./ton
Neunkirche Eisenwerke (p. 87)	1·41	17·2
Neunkirche Eisenwerke (p. 89)	1·35	19·8
Casselsteiner Eisenwerke (p. 89)	1·44	22·0

Since these are out of line with the general run of results, it would appear that they have had some

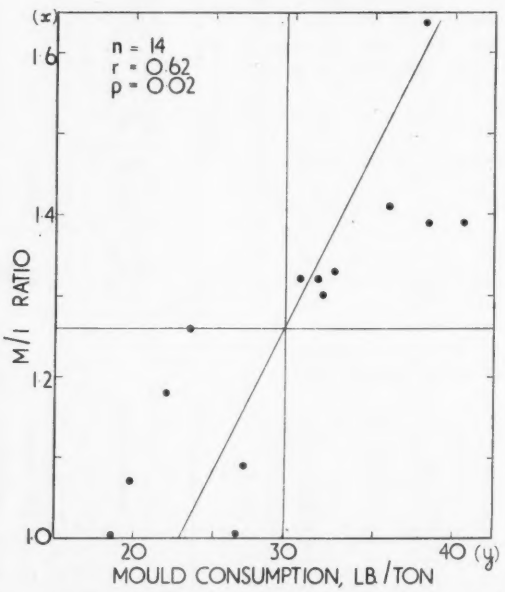


Fig. 11—Correlation between M/I ratio and mould consumption for wartime German square small-end-up moulds. Regression: 0·1 increase in M/I ratio increases mould consumption by 2·5 lb./ton

Table VI

BRITISH SQUARE SMALL-END-UP MOULDS 2½ TO 6 TONS IN WEIGHT

Mould No.	Mould Weight, tons	M/I Ratio	Consumption, lb./ton
I.M.S.C. 1st Report*			
A	3·1	1·00	24·7
J 1	3·0	1·34	33·0
N	3·2	0·98	24·5
1945 Questionnaire No. 1			
A 1	3·3	0·90	22·0
G 1	4·0	1·18	27·0
G 2	3·9	1·22	22·0
J 4	2·6	1·08	31·0
K 1–3	4·2	1·01	23·0
K 4–6	4·4	1·07	26·0
Mean	3·52	1·09	25·9

* Plates XXVIII and XXXI from Section VI of Iron and Steel Institute Special Report No. 16

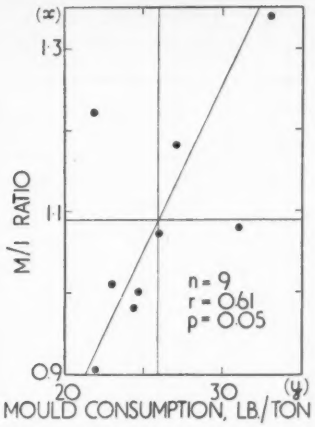


Fig. 12—Correlation between M/I ratio and mould consumption for British square small-end-up moulds. Regression: 0·1 increase in M/I ratio increases mould consumption by 2·38 lb./ton

exceptionally favourable treatment, possibly including the grinding of the inner faces during their working life.

Wartime German Moulds

It is interesting to compare these pre-war results with those obtained more recently and this has been done by taking relevant data for all the small-end-up square moulds of 2½ to 6 tons in weight from pp. 76–82 of the B.I.O.S. Report. These data are set out in Table V and Fig. 11. There is a very marked correlation between M/I ratio and consumption with no such backward trend above 1·25 M/I ratio as occurred in the case of the pre-war data. The regression is again very similar to that found in the other diagrams, *i.e.*, 2·5 lb./ton increase for each 0·1 increase in M/I ratio.

For comparison, the data for the British moulds of similar type, *i.e.*, small-end-up square moulds of 2½ to 6 tons have been abstracted from the I.M.S.C. First Report and the 1945 questionnaire, and are set out in Table VI and Fig. 12. As will be seen the regression is again in line with the other results, *i.e.*, 2·38 lb./ton for each 0·1 increase in M/I ratio.

A comparison of the British and German moulds of

Table VII

G.H.H. SLAB MOULDS (B.I.O.S. REPORT NO. 685, PAGE 80)

Mould Weight, tons	M/I Ratio	Mould Consumption, lb./ton
1·1	1·44	58·2
1·25	1·48	52·8
1·44	1·44	50·2
1·65	1·40	45·2
2·00	1·38	45·1
2·77	1·63	49·0
2·66	1·40	41·5
2·77	1·29	40·4
3·20	1·28	37·8
3·47	1·16	35·6
4·03	1·15	36·1
5·07	1·20	34·3
5·19	1·15	35·7
7·20	1·03	42·3
5·80	1·16	36·0
16·25	1·35	45·3
Mean 4·11	1·31	42·8

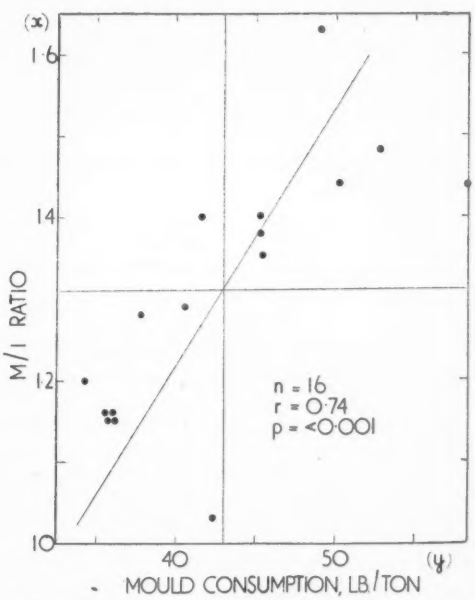


Fig. 13—Correlation between M/I ratio and mould consumption for G.H.H. slab moulds. Regression: 0·1 increase in M/I ratio increases mould consumption by 3·4 lb./ton

Table VIII

SMALL-END-UP SLAB MOULDS (OTHER THAN G.H.H.), BRITISH AND GERMAN

Mould Weight, tons	M/I Ratio	Mould Consumption, lb./ton
August Thyssen Hutte (from pp. 77–78 of B.I.O.S. Report)		
7·3	1·46	40·0
7·8	1·38	43·2
8·2	1·28	56·5
9·2	1·32	48·2
10·2	1·31	57·5
10·5	1·21	53·1
2·0	1·82	67·0
3·4	1·54	68·0
5·5	1·65	91·5
10·8	1·44	113·0
Ruhrstahl Henrichshutte (from p. 81 of B.I.O.S. Report)		
2·9	1·47	41·5
3·8	1·67	44·5
4·9	1·70	40·5
6·0	1·66	39·0
7·9	1·76	42·0
11·9	1·98	52·2
British (I.M.S.C. Questionnaire No. 1)		
4·5	1·00	43·0
6·1	0·86	26·0
7·6	1·20	52·0
7·3	1·15	60·0
21·2	1·25	62·0
British, Pre-War (I.M.S.C. First Report)		
6·5	1·18	26·0
8·1	1·18	34·8
6·1	0·84	45·0
8·2	0·82	42·5
8·9	0·89	25·7
7·2	1·08	44·0
Mean 7·55	1·34	50·3

similar type and used in the main over the same period is as follows:

	Number of Types	Mean M/I Ratio	Mean Consumption, lb./ton
British	9 (incl. 3 pre-war)	1·09	25·9
German	14	1·26	29·6

Slab Moulds Used by Gutehoffnungshutte (G.H.H.)

In the B.I.O.S. Report, on p. 80, the mean mould consumptions over a period of 4 years are given for 16 sizes of slab moulds ranging in weight from 1·1 to 16·25 tons. These data are considered separately, since they cover a period of four years for moulds used in one melting shop; as a result the other variables, *e.g.*, steel quality, method of pouring, stripping, cooling, etc., will have been of a more constant value than in the case of the mould data discussed earlier in this paper.

The M/I ratio, mould weight, and consumption of these moulds are set out in Table VII and the M/I ratio and consumption data are plotted in Fig. 13.

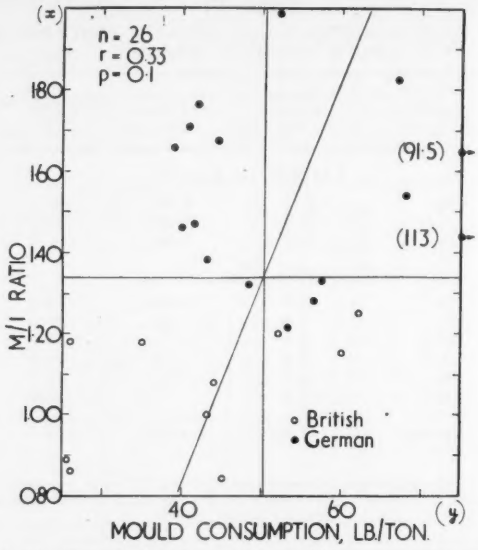


Fig. 14—Correlation between mould consumption and M/I ratio for British and German slab moulds. Regression: 0·1 increase in M/I ratio increases mould consumption by 2 lb./ton

It will be noted that the consumption increases at the rate of 3·4 lb./ton for each 0·1 increase in the M/I ratio. As might be expected, the scatter of the points is much less in Fig. 13 than, for example, in Figs. 1 and 2. It is interesting to note, also, that when the effect of the mould weight is eliminated (*see* Table III, item 5), the correlation between M/I ratio and mould

Table IX

OCTAGON MOULDS OVER 4 TONS IN WEIGHT

Mould Weight, tons	M/I Ratio	Mould Consumption, lb./ton
G 4·2	0·84	26
G 4·9	0·96	26
G 5·8	0·88	21
G 5·8	1·62	37
GK 8·2	0·94	58
B 8·7	1·08	45
GK 11·8	0·90	56
G 12·5	0·94	25
B 14·0	1·04	55
B 14·7	0·88	30
GK 15·5	0·89	56
GK 17·8	0·78	49
B 19·0	1·18	38
B 19·0	1·03	55
GK 19·8	0·68	36
B 22·9	1·12	53
G 33·1	1·06	22
G 34·7	0·68	25
GK 39·1	0·77	48
GK 46·5	0·64	47
B 51·5	0·84	35
GK 87·5	0·62	53
B 106·0	0·94	37
G 125·0	0·70	19
GK 160·0	0·61	53

G = German
GK = German (Krupps)
B = British

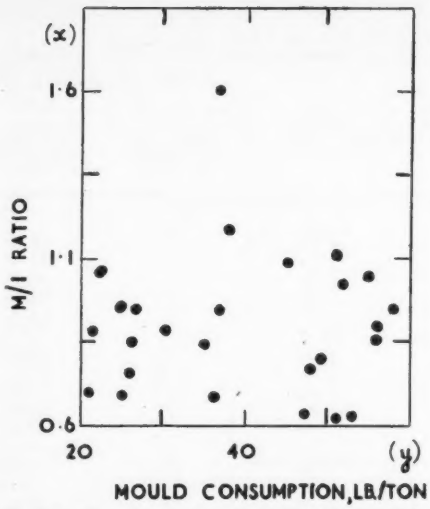


Fig. 15—Mould consumption plotted against M/I ratio for British and German octagon moulds

consumption has the very high value of 0·9, nearly perfect, whereas when mould weight is correlated against consumption with the effect of the M/I ratio eliminated, the degree of correlation is practically zero, *i.e.*, mould size within the weight range 1 to 16 tons has not affected mould consumption one way or the other.

COMBINED DATA FROM THREE SOURCES

Slab Moulds, other than G.H.H.

The data for the slab moulds other than G.H.H. have been abstracted from the three sources—the I.M.S.C. First Report, the 1945 questionnaire, and the B.I.O.S. Ingot Mould Report, and are presented in Table VIII; the mould consumption is plotted against the M/I ratio in Fig. 14.

The consumption increases with the M/I ratio at the rate of 2 lb./ton for each 0·1 increase in M/I ratio. Here again the mould size appears to have no effect

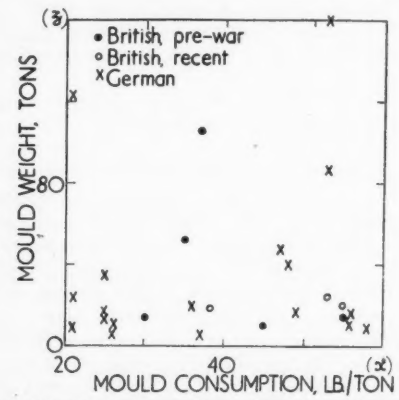


Fig. 16—Mould consumption plotted against mould weight for British and German octagon moulds

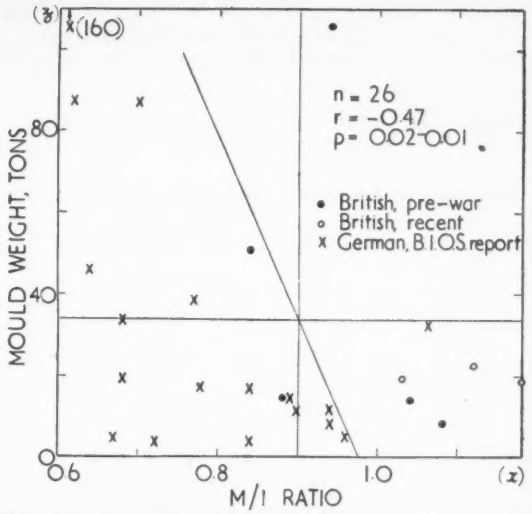


Fig. 17—Correlation between M/I ratio and mould weight for British and German octagon moulds. Regression: Each 1 ton increase in mould weight decreases M/I ratio by 0·002

on consumption when the M/I ratio effect is eliminated (*see* Table III, item 8).

Octagon Moulds

In Table IX the data for 25 types of octagon mould are given, taken from the three sources already mentioned. Neither mould weight nor M/I ratio bear any linear relationship to mould consumption, as is shown by Figs. 15 and 16. There is, however, a marked correlation between mould weight and M/I ratio (Fig. 17). As the mould gets bigger, the M/I ratio gets smaller, particularly when the moulds are over 30 tons in weight. It would appear probable from the following tables, that the optimum M/I ratio lies below 1·0 and that moulds above this value are likely to have a consumption rate above the average:

M/I Ratio	Number	Consumption, lb./ton
<0·8	7	39·6
0·8–1·0	10	37·0
>1·0	7	43·6

Or, expressing the results in another way:

Of 7 moulds with M/I ratio less than 0·8, 60% had a consumption above the average.

Of 7 moulds with M/I ratio above 1·0, 60% had a consumption above the average.

Of 10 moulds with M/I ratio 0·8–1·0, only 30% had a consumption above the average.

Comparison of British and German Octagon Moulds*

	Number	M/I Ratio	Consumption, lb./ton
British moulds ...	8	1·01	43·5
Krupp moulds ...	9	0·76	50·7
Other German moulds (excluding one of M/I Ratio 1·62) ...	8	0·86	23·6

The very good performance of the German moulds, other than Krupps, will be noted. None of the British moulds have a ratio less than 0·8; 3 are within the 0·8–1·0 range, and 5 above it, and the consumptions are 34·0 and 45·2 lb./ton respectively.

Table X

I.M.S.C. FIRST REPORT—MOULDS FOR INGOTS OF 50 CWT. OR MORE

Mould Number	Ingot Weight, cwt.*	Type of Mould†	Suggested M/I Ratio (see Fig. 18)	Actual M/I Ratio*	Mould Consumption, lb./ton	
					Within ± 0·1 of suggested M/I Ratio	Greater than + 0·1 over suggested M/I Ratio
A	63	SEU, SCT	1·02	1·00	24·7	...
B 6	160	BEU, Oct.	0·88	1·08	...	45·2
B 7	270	,, ,,	0·88	1·04	...	55·3
C 1	110	SEU, Slab	0·91	1·18	...	26·0
C 2	138	,, ,,	0·88	1·18	...	34·8
D	50	BEU, CB	1·05	1·51	...	53·7
E 2	66	,, ,,	1·01	0·98	35·3	...
E 3	80	,, Slab, CB	0·98	1·20	...	59·2
E 4	83	,, Sq., ,,	1·02	1·06	39·0	...
E 5	153	,, Slab, ,,	0·88	1·11	...	57·0
G 1	205	BEU, Oct.	0·88	1·56	...	56·0
G 2	330	,, ,,	0·88	1·36	...	52·6
G 3	535	,, ,,	0·88	1·34	...	57·2
H 2	70	,, Sq., CB	1·00	1·41	...	49·8
J 2	146	SEU, SCT, Slab	0·88	0·84	45·0	...
K 3	280	BEU, Oct.	0·88	1·05	...	30·2
K 4	1040	,, ,,	0·88	0·99	...	35·5
K 5	1960	,, ,,	0·88	1·08	...	37·2
M 1 (i)	200	SEU, Slab	0·88	0·82	42·5	...
M 1 (ii)	200	,, SCT, Slab	0·88	0·89	25·7	...
M 2 (i)	80	,, ,, ,,	0·98	1·00	29·5	...
M 2 (ii)	80	,, ,, ,,	0·98	1·00	29·5	...
N	65	SEU, Slab	1·01	0·98	24·5	...
R	55	,, Oct.	1·04	1·20	...	30·0
T 2	139	,, SCT, Slab	0·88	1·08	...	44·0
U	53	BEU, Sq., OE	1·04	0·95	57·0	...
W	66	SEU	1·01	1·37	...	30·5

* Excluding superimposed head (if any)

† SEU = Small end up
BEU = Big end up
CB = Closed bottom
SCT = Semi-closed top
OE = Open ended

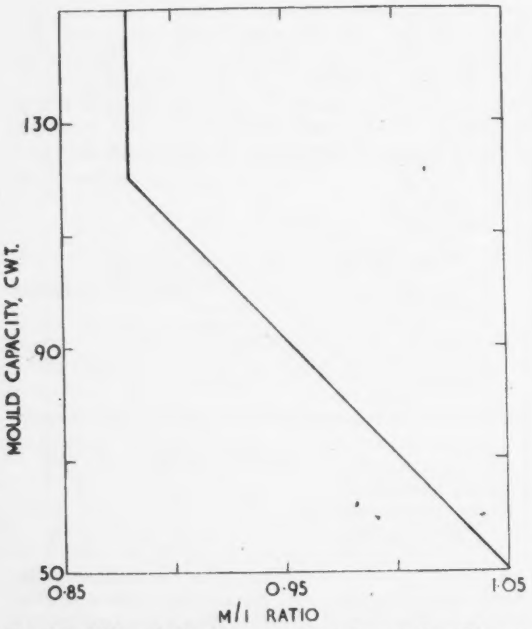


Fig. 18—M/I ratio for ingot moulds of more than 50 cwt. (excluding the weight of steel in any superimposed head)

OPTIMUM M/I RATIO

A consideration of the foregoing results suggests that the M/I ratio likely to give the lowest mould consumption lies between 0·8 and 1·0 for moulds of 6 tons capacity and upwards, and that for moulds below this capacity, the M/I ratio should increase as the mould capacity decreases. The suggested scale is shown in Fig. 18. The M/I ratio shown for any given capacity mould refers to the weight of the mould divided by the weight of steel in the mould itself, *i.e.*, the body or chill weight of the ingot, excluding the weight of the steel in any superimposed feeder head.

In the case of octagon moulds having superimposed feeder heads which contain for example, 20% of the total weight of steel, the M/I ratios would be:

Excluding steel in feeder head ... 0·88
Including steel in feeder head ... 0·704

The graph, Fig. 18, being based upon overall practice throughout the United Kingdom and Germany, is only useful as a general guide. Such factors as type of feeder head, if any, closed or open bottoms, and open or semi-closed tops, must be taken into consideration when the M/I ratio is being considered. For example, a 60 cwt. capacity mould open top and bottom might have an M/I ratio of 1·00 but a mould of the same capacity having the same mean wall thickness but a fully closed bottom would have

Table XI

I.M.S.C. 1945 QUESTIONNAIRE NO. 1—MOULDS FOR INGOTS OF 50 CWT. OR MORE

Mould Number	Ingot Weight, cwt.*	Type of Mould†	Suggested M/I Ratio (see Fig. 18)	Actual M/I Ratio	Mould Consumption, lb./ton	
					Within ± 0·1 of suggested M/I Ratio	Greater than + 0·1 over suggested M/I Ratio
A 1	78	SEU, SCT	0·99	0·86	22	...
A 2	78	SEU (Inserted Bricks at top)	0·98	0·88	28	...
A 3 and 4	303	BEU, Oct.	0·88	1·25	...	55
B 1–5	75	,, Sq., CB	0·99	1·34	...	37
B 6–8	102	,, Slab, ,,	0·92	1·52	...	90
B 9–12	56	,, Sq., ,,	1·04	1·78	...	73
B 14–17	74	,, Slab, ,,	0·99	1·22	...	43
C 4	250	,, Oct. ,,	0·88	1·52	...	38
D 1–6	165	SEU, Slab	0·88	1·13	...	44
D 3 (b)	165	,, ,,	0·88	1·13	...	45
F 2	267	BEU, Oct.	0·88	1·38	...	63
G 1	76	SEU, SCT	0·99	1·05	27	...
G 2	70	,, ,,	1·00	1·13	...	22
H	143	,, ,, Slab	0·88	0·86	26	...
J 3	50	BEU, Sq., OE	1·05	1·04	29	...
K 1–3	86	SEU, ,, ,,	0·96	0·98	23	...
K 4–6	86	,, ,, SCT	0·96	1·04	26	...
L 1	135	,, Slab	0·88	1·13	...	52
L 2	135	,, ,,	0·88	1·09	...	60
M	55	BEU, 12-sided, CB	1·04	1·20	...	55
N 1	320	,, Oct.	0·88	1·43	...	53
AB 1	341	SEU, SCT, Slab	0·88	1·24	...	62
AB 2	101	,, ,, Rect.	0·93	1·12	...	29
AB 3	122	,, ,, ,,	0·88	1·14	...	29
AB 4	124	BEU, Sq., OE	0·88	1·03	...	38

* Excluding superimposed head (if any)

† SEU = Small end up; BEU = Big end up; CB = Closed bottom; SCT = Semi-closed top; OE = Open ended

an M/I ratio of 1·13, owing to the additional weight of metal required to provide for the closed bottom.

Comparison of British Practice with Optimum M/I Ratio

In Tables X and XI the moulds used for casting ingots of 50 cwt. and more are tabulated, the data having been abstracted from the I.M.S.C. First Report and the 1945 questionnaire. In these tables the suggested M/I ratio is shown together with the actual M/I ratio. It will be noted that 17 of the moulds are within ± 0·1 of the suggested M/I ratio while 35 have an M/I ratio exceeding this range.

The mould consumptions are summarized in Table XII.

EFFECT OF CHANGE IN M/I RATIO ON MOULD CONSUMPTION

The effect of change in M/I ratio on mould consumption as shown by the data from the various

Table XII

COMPARISON OF MOULD CONSUMPTIONS

	Moulds within 0·1 of the Suggested M/I Ratio			Moulds with M/I Ratio More than 0·1 Greater than that Suggested		
	Number	M/I Ratio	Mould Consumption, lb./ton	Number	M/I Ratio	Mould Consumption, lb./ton
Small-end-up moulds:						
Square or nearly square ...	9	0·97	26·0	4	1·19	27·6
Slabs	4	0·86	34·8	8	1·14	46·0
Big-end-up moulds:						
Square or nearly square ...	4	1·01	40·1	6	1·36	51·4
Octagons*	...	...	...	12	1·26	48·2
Slabs	...	...	...	3	...	64·1

* Of these twelve octagon moulds, five are only just outside the proposed range, while the remainder are well outside it, *viz.*, the five moulds with M/I ratios of 1·08 or less have an average ratio of 1·05, and an average mould consumption of 40·7 lb./ton, and the seven moulds with M/I ratios of 1·25 or more have an average ratio of 1·39 and an average mould consumption of 53·5 lb./ton.

Table XIII

BIG-END-UP MOULDS FOR INGOTS 50 CWT. OR LESS TOTAL WEIGHT

Total Ingot Weight, cwt.	Mould Weight, cwt.	M/I Ratio	Life	Mould Consumption, lb./ton	Feeder Head*	Suggested M/I Ratio (from Fig. 19)
14	24	1·84	59	70·0	0	1·60
18	34	1·94	59	73·0	0	1·54
18	26	1·48	83	39·9	0	1·54
18	30	1·66	78	47·9	0	1·54
20	32	1·64	67	54·8	0	1·50
22	34	1·50	53	63·4	0	1·47
22	36	1·63	60	61·1	0	1·47
22	34	1·50	48	70·0	0	1·47
24	32	1·35	91	33·0	0	1·43
26	44	1·67	99	37·7	0	1·40
26	34	1·33	78	38·0	0	1·40
36	50	1·43	86	37·2	0	1·24
6	9	1·52	25	136·0†	1	...
8	12	1·51	30	113·0†	1	...
14	20	1·35	30	100·9†	1	1·65
14	26	2·00	73	61·0	1	1·65
16	22	1·38	30	103·6†	1	1·60
16	20	1·22	70	39·2	1	1·60
18	26	1·53	30	114·2†	1	1·57
18	30	1·70	30	127·3†	1	1·57
20	28	1·46	47	68·9	1	1·54
20	28	1·46	30	109·3†	1	1·54
20	40	1·98	30	147·8†	1	1·54
20	28	1·41	41	76·0	1	1·54
24	34	1·43	30	107·1†	1	1·48
24	32	1·33	50	59·7	1	1·48
26	34	1·35	67	45·0	1	1·45
26	40	1·60	27	133·0†	1	1·45
26	42	1·63	38	96·0†	1	1·45
28	40	1·48	30	110·6†	1	1·42
28	38	1·37	62	49·3	1	1·42
28	40	1·41	61	51·1	1	1·42
28	40	1·41	60	52·0	1	1·42
28	34	1·22	40	68·0	1	1·42
30	40	1·33	59	51·1	1	1·40
30	40	1·33	54	55·8	1	1·40
36	46	1·33	64	46·3	1	1·30
36	46	1·33	82	48·2	1	1·30
40	40	1·00	70	32·0	1	1·25
40	46	1·12	49	51·0	1	1·25
40	44	1·10	40	61·0	1	1·25
40	43	1·10	41	60·0	1	1·25
42	72	1·69	25	151·0†	1	1·22
42	46	1·09	38	64·0	1	1·22
42	46	1·09	26	94·0†	1	1·22
44	52	1·17	58	45·0	1	1·20
44	72	1·67	20	187·5†	1	1·20
44	50	1·15	12	214·5†	1	1·20
46	50	1·09	20	122·8†	1	1·17
46	60	1·30	40	72·7	1	1·17
50	44	0·86	20	96·8†	1	1·13
6	8	1·64	62	59·0	2	...
14	26	1·93	100	43·3	2	1·60
20	40	2·00	92	49·0	2	1·50
20	42	2·10	37	127·0†	2	1·50
34	48	1·46	32	102·0†	2	1·27
36	50	1·40	76	41·5	3	1·24
38	52	1·33	42	71·0	3	1·20
46	62	1·38	54	57·2	3	1·10
46	62	1·38	146	21·2	3	1·10

* 0 = None. 1 = Superimposed. 2 = Built in. 3 = Loose bricks.
† Excluded from the statistical examination.

sources may be summarized as follows:

	Increase in Consumption for Each Increase of 0·1 in the M/I Ratio, lb./ton
First Report of I.M.S.C. (45 types)	2·56
Questionnaire, 1945 (32 types) ...	5·45
Pre-war German moulds, up to M/I ratio of 1·25 (35 types) ...	2·80
G.H.H. slab moulds (16 types) ...	3·40
Other slab moulds, British and German (27 types)	2·00
British small-end-up square moulds (9 types)	2·38
German small-end-up square moulds (14 types)	2·50
Mean ...	3·01

MOULDS FOR INGOTS OF 50 CWT. OR LESS TOTAL WEIGHT

The foregoing examination was confined to moulds 50 cwt. or more in capacity. Data which have since become available in the replies to a further questionnaire issued by the I.M.S.C.* have enabled the examination to be extended to cover moulds down to 18 cwt. capacity.

In these replies a total of 60 items are listed, comprising information on big-end-up moulds used for ingots of 50 cwt. or less total weight. (*See* Table XIII.) Moulds for ingots without feeder heads and for those with superimposed heads were examined separately. A preliminary examination shows that 20 of the items refer to mould performances which are abnormal for some reason or other and so these have been excluded from the statistical examination.

Moulds for Ingots without Feeder Heads

The 12 moulds without feeder heads (Table XIII) have an average ingot weight of 22·2 cwt., an average M/I ratio of 1·58, and an average mould consumption of 52 lb./ton. They can be divided into two equal groups, *viz.*, those with an M/I ratio greater than the average, 1·58, and those with a smaller M/I ratio, as follows:

	Average Ingot Wt., cwt.	Average M/I Ratio	Average Mould Consumption, lb./ton
M/I ratio <1·58	24·6	1·43	47
M/I ratio >1·58	19·6	1·73	57

Although the lower M/I ratios give the better mould composition, the difference is not statistically significant.

Moulds with Superimposed Feeder Heads

A scrutiny of Table XIII shows that there is, overall, a marked difference in the M/I ratios of moulds with feeder heads between moulds for ingots weighing 36 cwt. or less, and moulds for ingots weighing 40–50 cwt. The two groups have therefore been examined separately.

Moulds for Ingots of 36 cwt. or Less—These moulds are listed in Table XIV. The average weight of ingot for the 14 moulds is 26 cwt., the average M/I ratio 1·39, and the average mould consumption 55 lb./ton. To make the M/I ratios comparable with those for moulds without feeder heads it is necessary to exclude the weight of metal contained in the superimposed heads. Assuming that the feeder head is 12% of the total weight, the average weight of ingot metal in the moulds is 88% of 26 cwt., *i.e.*, 22·8 cwt. This gives an average M/I ratio of 1·39 × 26 ÷ 22·8, *i.e.*, 1·58, which is exactly the same as the mean M/I ratio for the moulds without feeder heads (*see* above).

The 14 moulds may be divided into two groups—those with M/I ratio above the mean, and those with lower M/I ratio—as follows:

	Average Ingot Weight, cwt.	Average M/I Ratio	Average Mould Consumption, lb./ton
M/I ratio <1·39 (9 moulds):			
With feeder head ...	28·2	1·31	51
Excluding feeder head	24·8	1·49	58
M/I ratio >1·39 (5 moulds):			
With feeder head ...	22	1·53	62
Excluding feeder head	19·4	1·74	70

With this adjustment of the figures to allow for the metal in the superimposed feeder heads, it is possible

Table XIV

BIG-END-UP MOULDS WITH SUPERIMPOSED HEADS FOR INGOTS 36 CWT. OR LESS IN WEIGHT

Ingot Weight, cwt.	M/I Ratio	Mould Consumption, lb./ton
14	2·00	61
16	1·22	39
20	1·46	69
20	1·41	76
24	1·33	60
26	1·35	45
28	1·37	49
28	1·41	51
28	1·41	52
28	1·22	68
30	1·33	51
30	1·33	56
36	1·33	46
36	1·33	48
Mean 26	1·39	55

Table XV

COMPARISON OF MOULDS WITH AND WITHOUT FEEDER HEADS

M/I Ratio	Number of Moulds	Average Ingot Wt., cwt.	Average M/I Ratio	Average Mould Consumption, lb./ton
Moulds with Feeder Heads				
Below average ...	9	24·8	1·49	58
Above average ...	5	19·4	1·74	70
Moulds without Feeder Heads				
Below average ...	6	24·6	1·43	47
Above average ...	6	19·6	1·73	57

* Questionnaire No. 2, 1946.

Table XVI

MOULDS WITH SUPERIMPOSED HEADS FOR INGOTS OF 38–50 CWT.

Ingot Weight, cwt.	M/I Ratio	Mould Consumption, lb./ton
40	1·00	32
40	1·12	51
40	1·10	61
40	1·10	60
42	1·09	64
44	1·17	45
46	1·30	73
Mean of first 6 items:		
41	1·11	52
Mean of first 6 items excluding superimposed head :		
36	1·25	59

to make a fair comparison of the two types of mould with and without feeder heads (*see* Table XV).

It will be noted that the M/I ratios are almost identical and that the non-feeder-head moulds have the lower mould consumption, *i.e.*, moulds without feeder heads with M/I ratios below the average had a 19% lower mould consumption, and with M/I ratios above the average an 18·6% lower consumption, than moulds with feeder heads.

It would appear from the above figures that an M/I ratio of 1·45 is sufficient for moulds of about 25 cwt. capacity, and that ratios higher than this result in increased mould consumption of the order of 3–5 lb./ton for each increase of 0·1 in the M/I ratio. The above data alone are not sufficient to show a significant difference between the two groups, but the results are closely in line with those shown previously.

Moulds for Ingots of 38–50 cwt.—These moulds are listed in Table XVI. The mean of the first six items can be taken as a guide in suggesting a desirable M/I ratio for moulds within this range. Excluding the feeder head the figure is 1·25.

Moulds for Ingots of 14–24 cwt.—Most of these moulds are used without feeder heads. Details are given in Table XVII, in which the weight of steel in superimposed heads is excluded. Once again in general the lowest M/I ratios give the lowest mould consumptions.

OPTIMUM M/I RATIOS FOR MOULDS OF 18–50 CWT. CAPACITY

The optimum M/I ratios for moulds of 50 cwt. or more capacity, as deduced from the data, were shown in Fig. 18. It is now possible to extend the graph to cover the moulds down to about 18 cwt. capacity

Table XVII

MOULDS OF 14–24 CWT. CAPACITY
(Excluding the weight of steel in any superimposed heads)

Mould Capacity, cwt.	Actual M/I Ratio	M/I Ratio (from Fig. 19)	Mould Consumption, lb./ton	Source of Data
14	1·84	1·70	70	First 9 items of Table XIII
18	1·94	1·60	73	
18	1·48	1·60	40	
18	1·66	1·60	48	
20	1·64	1·53	55	
22	1·50	1·50	63	
22	1·63	1·50	61	
22	1·50	1·50	70	
24	1·35	1·43	33	
14	1·38	1·70	45	Items 2 to 6 of Table XIV*
18	1·60	1·60	78	
18	1·60	1·60	86	
21	1·51	1·51	68	
23	1·54	1·45	51	
14	1·93	1·60	43	Table XIII, 7th and 8th items from bottom
20	2·00	1·50	49	
Mean 19·1	1·63		58·3	

	Number	Ingot Weight, cwt.	M/I Ratio	Mould Consumption, lb./ton
Moulds with M/I $\leq$ 1·60	7	19·5	1·47	53
,, ,, ,, $\geq$,, ...	9	18·0	1·76	63

* Excluding weight of metal in the superimposed feeder heads

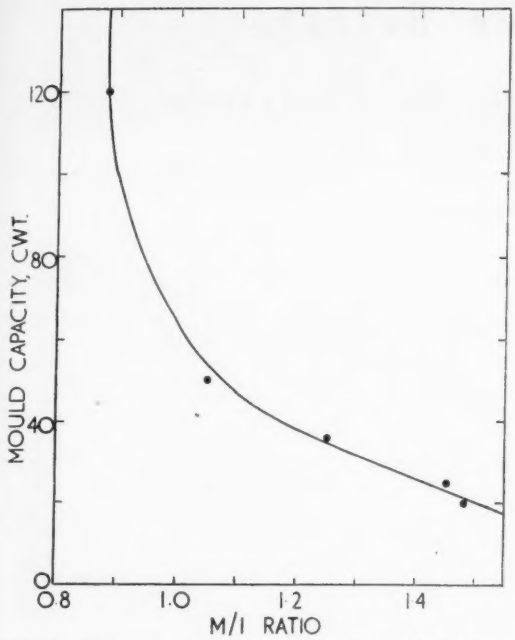


Fig. 19—Optimum M/I ratio for ingot moulds of more than 18 cwt. (excluding the weight of steel in any superimposed head)

(excluding the weight of steel in superimposed feeder heads), using the following points :

Ingot Weight, cwt.	M/I Ratio
36	1·25
25	1·45
20	1·50

The complete curve is shown in Fig. 19.

It should be pointed out that the evidence for the last three points is less strong than that for the remainder of the curve, and that, although these points are below the average M/I ratios at present used, the true optimum values may in fact be lower still. It is, however, evident from the data that the M/I ratio is of much less importance for moulds of less than 50 cwt. capacity than for moulds of more than 50 cwt. capacity.

REASONS FOR THE USE OF HIGH M/I RATIOS

The reasons why so many moulds have been designed with M/I ratio in excess, and sometimes greatly in excess, of what appears to be desirable, are probably as follows :

(*a*) The rate of solidification of ingots is thought by some to be increased by very thick mould walls.

(*b*) Moulds must be heavy enough to withstand rough usage, *e.g.*, inverting and bumping to strip the ingot.

(*c*) Moulds are increased in thickness in order to overcome major cracking failures.

Results obtained recently show that some mould metals are more resistant to major cracking than others. It would appear that, in general, moulds have been thickened in an effort to overcome previous failure through cracking, although such failures are primarily due to weak metal, poor design, or both.

CALCULATION OF MOULD WALL THICKNESS REQUIRED

The mean wall thickness required for any given M/I ratio shown in Fig. 19 is given by the equation :

$$\text{Wall thickness} = 0{\cdot}21\ \text{M/I}\sqrt{\text{A}}\ \text{in.}$$

for round, octagon, or nearly square moulds, where M/I is the M/I ratio and A is the cross-sectional area in square inches.

For example, a 24-in. octagonal mould of 4 tons capacity requires a wall thickness of

$$0{\cdot}21 \times 0{\cdot}94 \times \sqrt{504} = 4{\cdot}43\ \text{in.}$$

For slab moulds of increasing broad to narrow side ratio (assuming walls of equal thickness), the equation becomes :

$$\text{Wall thickness} = 0{\cdot}20\ \text{M/I}\sqrt{\text{A}}$$

for a broad to narrow side ratio of 2·0.

CONCLUSIONS

(1) For a mould of any given capacity, there is an optimum weight, beyond which additional metal serves only to increase the mould consumption, in proportion to the additional weight.

(2) Good design is an indispensable condition for the achievement of maximum mould economy. The incorporation of the important factor, optimum M/I ratio, into the design, will fail in its purpose if the mould is ill-designed in respect of other features.

(3) Efforts to offset poor design, poor foundry technique, or weak metal, by increasing the mould wall thickness appear from the data to have failed in their purpose.

Acknowledgments

The author is indebted to the members of the Ingot Moulds Sub-Committee for the interest they have taken in this work and for their helpful criticism, and particularly to Mr. R. H. Myers and Mr. W. H. Glaisher of B.I.S.R.A. for the trouble they have taken in reviewing the paper before publication; also to the many firms for the care with which they filled in the three questionnaires and without whose co-operation such an investigation would not have been possible.

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Ingot Surface Defects

FORMATION OF "DOUBLE SKIN" OR "CURTAINING" ON TOP-POURED MILD-STEEL INGOTS

By P. Walker

SYNOPSIS

This paper describes work carried out to determine the cause of the surface defect present on top-poured mild-steel ingots and known as "double skin" or "curtaining." The results indicate that two factors are of significance, (*a*) trapping of deoxidation slag near the ingot surface during solidification and (*b*) splash moving forward from the mould wall and allowing liquid steel to flow behind it. The former factor gives a "double skin" effect not visible on the surface, and the latter a visible "double skin" or "curtaining."

SUBSEQUENT to discussions in committee, on the nomenclature of ingot defects, an investigation has been made into the formation of the defect which it was agreed to call "double skin," and of which Fig. 1 is an illustration. The extent of this type of defect has been seen to vary in size from a

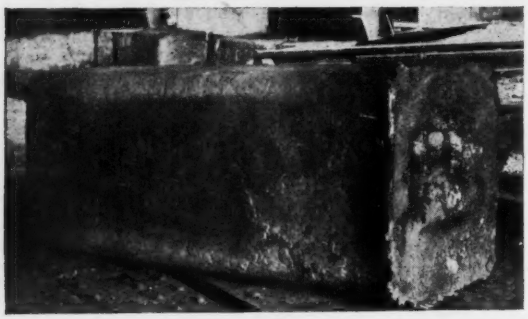

Fig. 1—Ingot showing "curtaining" or "double skin"

patch 1 in. in dia. to areas covering a considerable part of one or more of the ingot surfaces, as shown in Fig. 2.

Whilst the explanation of the defect having been caused by liquid metal exuding from a crack in the ingot surface may be satisfactory for the smaller patches, it cannot be accepted for the larger areas, such as those depicted in Fig. 2.

It became apparent at the outset of the investigation that the term "double skin" did not fully describe the defect, since in all the samples examined the area of double skin only extended to about $\frac{1}{4}$ in., this being confined to the area of junction of the two surfaces. It was noted, however, that a further type of "double skin" did exist lower down the ingot. This is shown diagrammatically in Fig. 3.

Paper SM/BD/62/47 of the Sub-Committee on Ingot Surface Defects of the Steelmaking Division of the British Iron and Steel Research Association. The opinions expressed in it are the author's, and are not necessarily endorsed by the Committee as a whole.

Manuscript received 19th August, 1947.

Mr. P. Walker is Chief Metallurgist of the Cargo Fleet Iron Co., Ltd.

Large Areas of "Curtaining" or "Double Skin"

Figure 4 is a macrograph showing typical curtaining and double skin. The junction between the raised and lower ingot surfaces can be seen in greater detail in Fig. 5. The position and contour of the larger areas of the defect gave rise to the opinion that the splash or surging which occurs on opening the ladle stopper may be responsible for the defect. It is suggested that the splash first formed on the mould wall as the result of the stream of metal hitting the mould stool,

Fig. 2—Diagram to illustrate extent of "double skin"

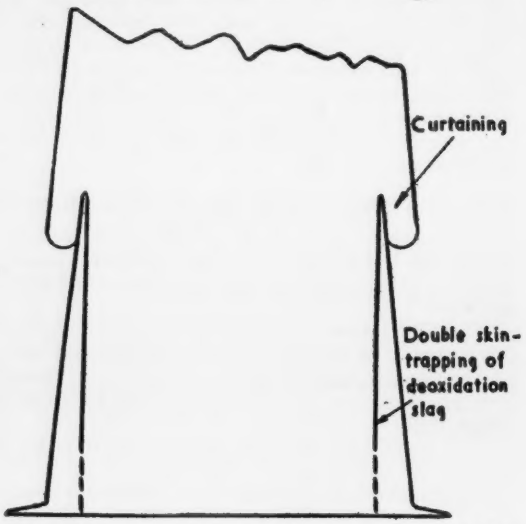

Fig. 3—Diagram to illustrate the two types of "double skin"

Fig. 4—Macrograph of cross-section of " curtaining " or " double skin." × ½

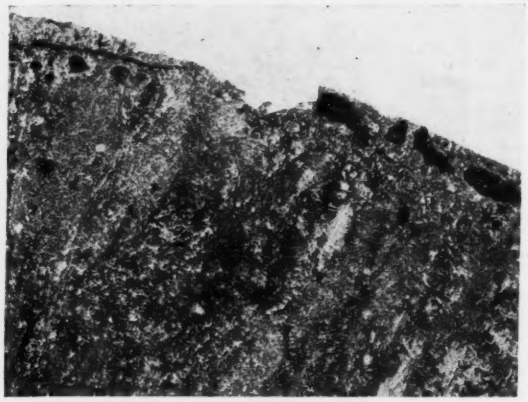

Fig. 5—Macrograph of cross-section of defect, showing junctions between raised and lower ingot surfaces. ×2½

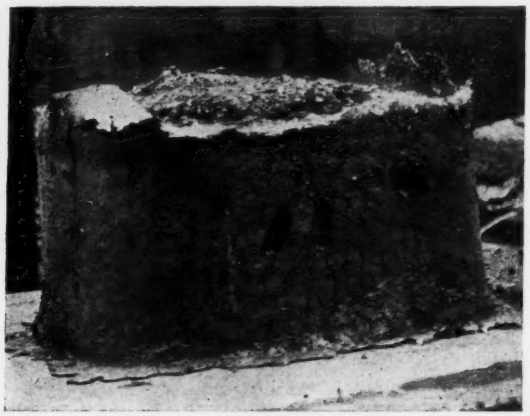

Fig. 6—" Box " filled with steel

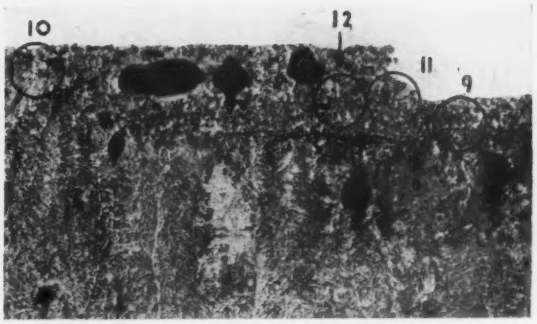

Fig. 8—Macrograph showing the position of the micrographs in Figs. 9, 10, 11, and 12. × 4

chills, and on withdrawing leaves a space into which the rising metal can flow to a limited extent. The lower portion of the ingot surface is the splash produced during " opening out " the stopper, and the raised portion is the surface produced by pouring after the well of metal has been formed with only a small amount of splashing taking place.

In order to obtain an idea of the extent of the initial splash a number of moulds into which the stopper had been opened out only momentarily, *e.g.*, ingot butts, were examined. Figure 6 shows a "box" filled with steel, whilst the photographs of Fig. 7 illustrate the effect produced, the splash reaching heights varying from 12 to 36 in. from the bottom of the ingot, the thickness of the splash being ½ in. at some points. The cooling of such splash will result in it shrinking from the mould wall and the rising metal will eventually overflow to some extent into the gap produced ; or alternatively liquid metal may flow through cracks in the splash.

The micro-examination of samples cut from the positions shown in Fig. 8 showed the presence of oxide and deoxidation slag trapped within the lower surface of the ingot, but not within the raised surface of the ingot. Also of note is the degree of decarburization present on the lower surface as compared with the raised surface. These two features are illustrated in Figs. 9 and 10. The junction between the initial splash and the metal which has flowed down the contraction gap is shown in Figs. 11 and 12 ; the presence of oxide at this point is to be noted.

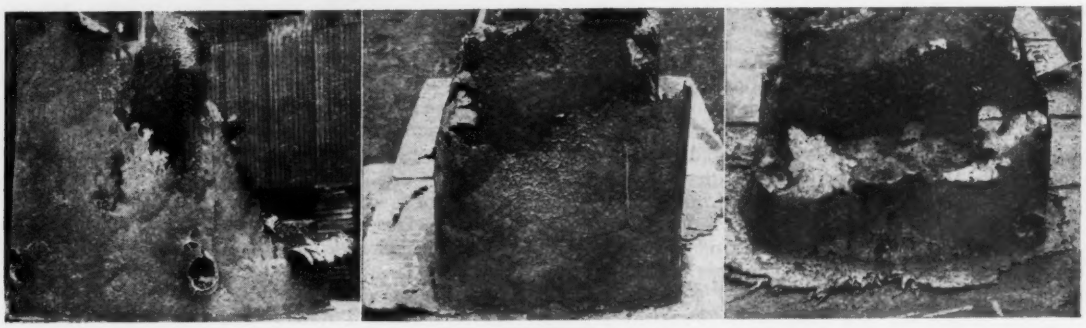

Fig. 7—" Boxes " produced by splash

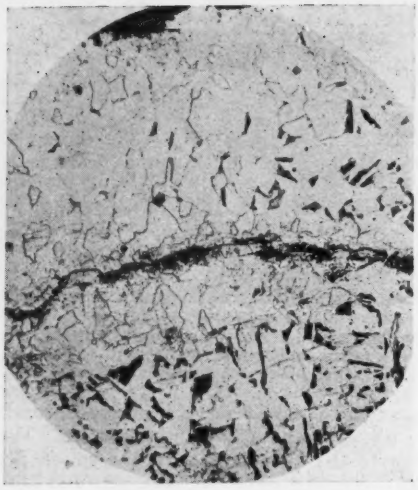

Fig. 9—Micrograph showing oxide and deoxidation slag trapped below the lower ingot surface, and decarburization of the ingot skin. Etched. × 30

Fig. 10—Micrograph showing structure immediately below the raised portion of the ingot skin. (*Cf.* the decarburization with that in Fig. 9.) Etched. × 30

Fig. 11—Micrograph showing junction between splash and metal which has flowed into the contraction gap. Etched. × 30

Fig. 12—Micrograph showing continuation of junction in Fig. 11. Etched. × 30

During the course of an earlier experiment in which two ingots were poured, one into a tarred mould and one into an untarred mould, a notable difference between the two ingots was the absence of curtaining on the ingot poured into the tarred mould. It is concluded that the splash did not adhere to the tarred mould wall, and was immediately taken up by the liquid metal.

It has also been noted that ingots which were poured into badly crazed moulds did not show curtaining. The absence of the defect on such ingots is concluded to be due to the splash "keying" itself on to the mould wall, and thus not producing a contraction gap into which metal can flow.

Acknowledgments

The author desires to thank the management of the Cargo Fleet Iron Co., Ltd., for permission to publish this paper.

The Protection of Steel by Various Non-Metallic Coatings

By J. C. Hudson, D.Sc., and T. A. Banfield, Ph.D., F.R.I.C.

SYNOPSIS

This is an interim report on the behaviour of a range of protective non-metallic coatings applied to structural mild steel and exposed to field corrosion tests as part of the investigations conducted by the Protective Coatings Sub-Committee of the British Iron and Steel Research Association. The atmospheric-exposure tests cover periods of up to 5 years, and the coatings tested can be grouped under (1) painting schemes, (2) sprayed coatings, and (3) other coatings. The results to date may be summarized as follows.

(1) Two-coat painting schemes applied to pickled steel were in good condition after 5 years, but when one of these schemes was applied to steel prepared by weathering and wire-brushing, failure was observed after 1½ years. The use of a proprietary inhibitive surface wash containing phosphoric and chromic acids on the partially descaled surface resulting from the latter treatment failed to improve matters. Lanolin-base primers incorporating chromate pigments and hardened with synthetic resin have given good results when applied to pickled steel.

(2) A process in which paint containing no volatile matter is sprayed-on hot has given good results at Sheffield but the coating shows signs of breakdown after 5 years' exposure in non-industrial atmospheres. No sign of corrosion of the encased steel was visible after 5 years in the case of specimens coated with cement/asbestos to a thickness of ¼ in.

(3) A tar/tallow/lime mixture applied hot protected the steel for only about 3 years in an industrial atmosphere, but this degree of protection is comparable with that to be expected from a single paint coat of approximately the same weight. A rubber/wax sheathing applied by wrapping has given good results in an industrial atmosphere, but less favourable ones in a marine atmosphere. Vitreous enamel coatings were in perfect condition, apart from minor mechanical damage, at all the exposure stations after 5 years.

The sea-water exposure tests showed that heavy coatings of sprayed bitumen, alone or mixed with rubber or zinc, gave fairly good results over a period of 2 years' continuous immersion. The addition of glass to a sprayed zinc coating had no appreciable effect. Vitreous-enamel coatings were not appreciably affected after 2 years, apart from slight damage.

Introduction

In a recent paper,[1] the authors have given an interim report on the results of tests by the Protective Coatings (Corrosion) Sub-Committee on protective metallic coatings suitable for use on structural steel. The purpose of the present communication is to present a similar account of the progress of a parallel investigation, commenced at the same time, of other protective schemes, comprising both paints and other types of coatings of a non-metallic character. As in the case of the metallic coatings, arrangements were made for tests both under conditions of atmospheric exposure and of complete immersion in sea-water.

Atmospheric-exposure tests are being conducted in several localities both at home and overseas. In this country they were commenced during 1940, but exposure abroad took place a year or so later, in 1941 or 1942. The total-immersion tests in sea-water were begun in 1941 and completed in 1943, when the specimens were removed for final examination. It will thus be seen that the results presented here relate to periods of 5 years' exposure or less in the case of the atmospheric tests and to 2 years' immersion in the case of the tests in sea-water. The former are being continued and it is hoped that further reports on their progress will appear later.

If tests of this type are to have their maximum practical value, it is essential that, wherever possible, the protective schemes shall be applied under industrial conditions at works where the various processes are in operation. In the case of the present investigation, this procedure was rendered possible by the willing co-operation of several firms and research institutions invited to help with the tests, not only in applying the protective schemes but also in permitting the investigators to visit their works or laboratories to record details of the procedures followed. The Sub-Committee's thanks for assistance in this manner, or for the provision of exposure facilities, are due to the following individuals and organizations:

The Air Ministry.
Mr. A. Barlow, A.M.Inst.C.E., Borough Engineer, Gosport.
British Insulated Cables, Ltd.
Great Western Railway Company.
London, Midland and Scottish Railway Company.
Marine Department, Lagos, Nigeria.
Mr. A. Pahl.
Parkinson Stove Co., Ltd.
Prodorite, Ltd.
Messrs. J. W. Roberts, Ltd.
Schori Metallising Process, Ltd.
South African Railways and Harbours.
Dr. H. Sutton.
Whessoe Foundry and Engineering Co., Ltd.

Paper No. MG/BA/15/47, submitted by the Protective Coatings Sub-Committee of the Corrosion Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association. The views expressed are the authors', and are not necessarily endorsed by the Sub-Committee as a body.

The manuscript was received on 21st August, 1947.

Dr. J. C. Hudson is Head of the British Iron and Steel Research Association's Corrosion Laboratory, Birmingham, and is Official Investigator to the Corrosion Committee. Dr. T. A. Banfield was formerly Assistant Investigator to the Corrosion Committee.

In addition, the authors wish to add their personal thanks to the Chairman, Mr. T. M. Herbert, and members of the Protective Coatings (Corrosion) Sub-Committee, for advice and encouragement as well as for practical assistance.

EXPERIMENTAL DETAILS

An outline of the experimental programme, showing the type of specimens, protective schemes tested and exposure stations is given in Table I. Further details of the protective schemes and of the methods of preparing the specimens will be found in the Appendix.

(*a*) Specimens

The standard specimen used in this investigation was a mild-steel flat measuring 15 × 10 × $\frac{3}{8}$ in.

(*b*) Protective Schemes

The protective schemes tested have been classed in three groups, namely, paints, sprayed coatings, and other coatings.

The painting schemes include two- and four-coat combinations of red lead in linseed oil and red oxide in linseed oil, a proprietary scheme of paints bound with alkyd-resin media, and painting schemes incorporating priming coats of resin-reinforced lanolin-base paints. These were applied in the laboratory to specimens prepared for painting by a number of different surface treatments.

The sprayed coatings comprise a paint with no volatile constituent, prepared for spraying-on hot by a special pistol, cement/asbestos as used chiefly for insulation purposes, powdered bitumen mixtures, and a mixture of powdered zinc with a high-silica glass, these powder mixtures being sprayed by a modified form of the powder metal-spraying pistol.

The other coatings tested are a tar/tallow/lime mixture applied hot, vitreous enamel, and a rubber/wax mixture which can be moulded to give a continuous sheath.

(*c*) Examination of the Coated Specimens before Exposure

The weight of coating was determined by weighing each specimen before and after coating, and notes were made on the appearance of the coatings, positions of defects, etc.

(*d*) Exposure Stations

Particulars of the climatic conditions at the five* atmospheric-exposure stations will be found in Table II. At Gosport the specimens were immersed in a vertical position beneath the pontoons of the landing stage, whilst at Caernarvon immersion took place from the raft of the Marine Corrosion Sub-Committee moored in the Menai Straits, the specimens being immersed vertically at a depth of from 18 to 24 in.

In the atmospheric-exposure tests the specimens were suspended in a vertical position by means of hooks of steel wire of $\frac{1}{4}$-in. dia., so as to be freely exposed.† The hooks were sherardized and dip-painted; this treatment has protected them satisfactorily so far, and no trouble from rust-staining of the specimens by rain-water running down from the hooks has occurred.

(*e*) Inspection of the Specimen and Final Examination

(i) *Field Inspections*—The atmospheric specimens have been inspected regularly. At the home stations

* The specimens at Singapore were lost after reports covering the first 6 months' exposure had been received, so no further reference will be made to this station.

† Details of this method of exposure will be found in the authors' earlier paper (*see* reference 1).

Table I

SCHEDULE OF TESTS ON PAINTS AND OTHER NON-METALLIC COATINGS

Basis Specimen	15 × 10 × $\frac{3}{8}$-in. Mild-Steel Flats (C 0·2%, Mn 0·6%, Si 0·05%)		
Coatings	*A*—Painting Schemes	*B*—Sprayed Coatings	*C*—Other Coatings
	(1) Over steel pickled by the sulphuric-acid/phosphoric-acid process (2) Over weathered and wire-brushed steel (3) Lanolin-base primers	(4) Hot-spray paint (5) Cement/asbestos (6) Bitumen mixtures (7) Powdered zinc and high-silica glass	(8) Tar/tallow/lime applied hot (9) Vitreous enamel (10) Rubber/wax sheathing
Exposure Stations	Atmospheric		Marine
	Home	Overseas	
	Calshot Llanwrtyd Wells Sheffield	Apapa (Nigeria) Congella (S. Africa) Singapore	Caernarvon Gosport

all the observations have been made by the same investigator, whilst at the overseas stations the Sub-Committee is indebted to local observers for assistance in this respect. Records are taken of blistering and peeling of the coating, and visual estimates are made of the percentage of the surface that is (*a*) discoloured, (*b*) rust-stained, or (*c*) rusted. The general distribution of these faults is noted, and also the condition of the edges of the specimens if different from that of the main surfaces. In all cases individual records are taken for the front and back surfaces. Any breakdown that is obviously due to proximity to the edges or the suspension holes, or to accidental damage, is noted separately. These reports are entered on standard forms which enable the progress of the corrosion of each specimen to be followed.

On the basis of the reports, the conditions of the front and back surfaces of each specimen at each inspection are graded according to the following scale:

x Excellent; no sign of breakdown of the coating.
f Fair; coating showing signs of breakdown, *e.g.*, blistered, discoloured, or rust-stained to any extent, or the underlying steel rusted over less than 5% of the area.
o Failed; 5% or more of the steel rusted.

The life of the protective coating on each surface is deduced from the gradings made at successive inspections. Two figures are computed, namely, *Lf*, the life to grade "fair," *i.e.*, the life to the first signs of breakdown, and *Lo*, the life to grade "failed," *i.e.*, the life to 5% of the surface rusted. The use of these figures for the lives of the coatings enables comparisons between different coatings to be readily made.

The results of immersion tests in the sea at Caernarvon have already been published[2]; in these particular tests the specimens were inspected at intervals of 4–6 weeks. At Gosport, however, the conditions of immersion were such that inspection *in situ* was not possible and no observations were made until the specimens were removed after 2 years' immersion. They were then scraped free from marine growths, washed, and returned to the laboratory for examination.

(ii) *Final Laboratory Examination after Removal*—At the atmospheric-exposure stations in this country, specimens graded as "failed" on the front and back surfaces at two or more consecutive inspections are removed as convenient. In the laboratory they are washed and weighed, and the coatings are then examined in detail, careful estimates of the percentage of surface rusted, etc., being made, as in the field inspections. The sea-water specimens were examined in the same way, after exposure.

EXPERIMENTAL RESULTS

An interim statement of the results of the atmospheric-exposure tests is shown in Table III; this covers a period of 5 years' exposure in the case of the home stations and shorter periods overseas. The results are classified under the exposure station, and the following figures are given for each specimen:

W Weight of coating, grammes per specimen.*
Lf Life of coating to grade "fair," years. Mean value for the front and back surfaces.
Lo Life of coating to grade "failed," years. Mean value for the front and back surfaces.

The figures for *W* were determined by weighing the specimen before and after coating (*see* Appendix); in the case of paint coatings the total weight of dry paint is given. The figures for *Lf* and *Lo* were deduced from the results of the field inspections, as already described. Laboratory examination of the specimens removed confirmed the field inspections, but it was found that determinations of the changes in weight on exposure were of little assistance for comparing different protective schemes.

The results of the total-immersion tests at Gosport, which it will be recalled relate to a period of two years, will be found in Table IV. These are based on

* The coating weights in grammes per specimen may be converted into ounces per square foot by multiplying by 0·016.

Table II

PARTICULARS OF THE ATMOSPHERIC-EXPOSURE STATIONS

Exposure Stations	Approximate Position		Average Meteorological Data			Observed Rates of Corrosion, mils/year			Date of Exposing Coated Specimens
	Latitude	Longitude	Temperature, °F.	Rainfall, in./year	Relative Humidity, %	Standard Mild Steel X*	Ingot Iron†	Pure Zinc‡	
Home									
Sheffield	53° N.	1° W.	48	30	84	4·21	4·95	0·64	16.7.40
Llanwrtyd Wells ...	52° N.	4° W.	47	55	79	1·24	2·37	0·12	6.8.40
Calshot	51° N.	1° W.	51	26	84	1·62	2·86	0·14	27.8.40
Overseas									
Congella, S. Africa ...	30° S.	31° E.	71	43	76	1·84§	3·27§	0·16§	8.4.42
Apapa, Nigeria ...	7° N.	4° E.	80	72	79	0·77	1·09	0·04	20.3.41

* Results of 5-year tests on 15 × 10 × ⅛-in. specimens exposed with mill scale.
† Results of 1-year tests on 4 × 2 × ⅛-in. pickled specimens, changed annually. Mean value for the years during which the coated specimens have been exposed.
‡ Results of 1-year tests on 4 × 2 × $\frac{1}{20}$-in. specimens, changed annually. Mean value for the years during which the coated specimens have been exposed.
§ These results were obtained at a site used previously at Congella, about ½ mile from the present site. The figures for ingot iron and zinc are the means of nine observations.

the laboratory examination of the specimens after removal, since, as already stated, no inspection in the field was possible.

DISCUSSION OF THE RESULTS OF THE ATMOSPHERIC-EXPOSURE TESTS

(1) PAINTING SCHEMES

Details of the formulations of most of the paints referred to in this sub-section will be found in the Appendix.

(*a*) Over Steel Pickled by the Sulphuric-Acid/Phosphoric-Acid Process

The method of surface preparation by sulphuric-acid/phosphoric-acid pickling is a good one, and it is unusual for paint to give poor results on specimens so prepared. In conformity with this, none of the five different painting schemes included in this section had reached the point of failure after 5 years' exposure, although there were signs of breakdown. The appearance of some of these painting schemes at this stage may be illustrated by the photographs reproduced in Figs. 1 and 2.* These show, at actual size, parts of the front surfaces of specimens exposed at Llanwrtyd Wells or Calshot. Signs of failure are evident on each specimen but, on the whole, these were less marked for painting schemes where a red-lead priming paint was used than for other schemes where red lead was not used. It should be noted (*see* Appendix) that the latter were all two-coat painting schemes, as compared with the four or five coats of the schemes based on red-lead primers; they were therefore at a considerable disadvantage as regards the important factor of total dry-paint film thickness.

* These photographs were taken after 5·6 years' exposure in both cases, *i.e.*, a little later than the end of the 5-year period covered by this interim report.

Table III

RESULTS OF THE ATMOSPHERIC-EXPOSURE TESTS ON PAINTS AND OTHER NON-METALLIC COATINGS

Coating Process*	Exposure Stations														
	Sheffield			Llanwrtyd Wells			Calshot			Congella, S. Africa			Apapa, Nigeria		
	W	*Lf*	*Lo*	*W*	*Lf*	*Lo*	*W*	*Lf*	*Lo*	*W*	*Lf*	*Lo*	*W*	*Lf*	*Lo*
(*A*) PAINTING SCHEMES															
1. Over Steel Pickled by the Sulphuric-Acid/Phosphoric Acid Process															
(*a*) Red oxide (2 coats)	21	3·0	>5·0	23	1·5	>4·9	22	1·6	>5·2	..	..	..	..	..	..
	22	3·0	>5·0	22	1·4	>4·9	21	1·0	>5·2	..	..	..	..	..	..
(*b*) Red lead (1 coat), red oxide (3 coats)	44	4·0	>5·0	46	2·0	>4·9	45	3·3	>5·2	..	..	..	..	..	..
	43	3·0	>5·0	47	2·0	>4·9	44	2·5	>5·2	..	..	..	..	..	..
(*c*) Red lead (2 coats), red oxide (3 coats)	60	4·0	>5·0	58	2·0	>4·9	62	3·0	>5·2	..	..	..	59	2·3	>4·6
	59	3·0	>5·0	61	2·0	>4·9	57	2·5	>5·2	..	..	..	58	2·3	>4·6
(*d*) Proprietary paints bound with alkyd-resin media (2 coats)	11	3·0	>5·0	11	1·0	>4·8	..	..	..	..	..	..	..	..	..
	12	3·0	>5·0	12	0·6	>4·8	..	..	..	..	..	..	..	..	..
(*e*) Inhibitive wash followed by a 2-coat proprietary scheme	13	1·6	>5·0	..	..	..	..	..	..	..	..	..	..	..	..
	11	1 6	>5·0	..	..	..	..	..	..	..	..	..	..	..	..
2. Over Weathered and Wire-Brushed Steel															
(*a*) Painted as in process 1 (*e*) above	14	1·0	1·6	15	0·6	1·0	13	0·3	1·0	..	..	..	..	..	..
	14	1·0	1·6	14	0·3	0·6	14	0·3	2·3	..	..	..	..	..	..
3. Over Steel Pickled with Sulphuric Acid															
(*a*) Lanolin primer *J* (2 coats), red oxide (2 coats)	39	5·0	>5·0	39	2·5	4·2	..	..	..	38	1·0	>3·0	..	..	..
	39	5·0	>5·0	37	2·5	4·2	..	..	..	40	1·0	>3·0	..	..	..
(*b*) Lanolin primer *K* (2 coats), red oxide (2 coats)	44	5·0	>5·0	44	2·5	4·5	..	..	..	42	1·0	>3·0	..	..	..
	44	5·0	>5·0	43	2·5	4·7	..	..	..	42	2·1	>3·0	..	..	..
(*c*) Red lead (2 coats), red oxide (2 coats)	58	5·0	>5·0	57	2·9	>4·8	..	..	..	61	1·0	>3·0	..	..	..
	55	5·0	>5·0	61	2·7	>4·8	..	..	..	57	2·5	>3·0	..	..	..
(*B*) SPRAYED COATINGS															
4. Hot-Spray Paint	41	>5·0	>5·0	39	3·3	>4·9	24	1·0	>4·1	..	..	..	..	..	..
	32	>5·0	>5·0	25	0·6	>4·8	37	0·5	>5·2	..	..	..	..	..	..
5. Cement/Asbestos Over:															
(*a*) Steel pickled by the duplex sulphuric-acid/phosphoric-acid process	1456	>5·0	>5·0	1268	4·6	>4·9	1434	>5·2	>5·2	..	..	..	..	..	..
	1252	>5·0	>5·0	1439	4·6	>4·9	1163	>5·2	>5·2	..	..	..	..	..	..
(*b*) As (*a*), with a priming coat of red-lead paint	1270	>5·0	>5·0	1294	4·6	>4·9	1480	>5·2	>5·2	..	..	..	..	..	..
	1389	>5·0	>5·0	1415	4·6	>4·9	1309	>5·2	>5·2	..	..	..	..	..	..
(*c*) As-rolled steel	1358	>5·0	>5·0	1368	3·8	>4·8	..	..	..	..	..	..	..	..	..
	1290	>5·0	>5·0	1389	4·6	>4·9	..	..	..	..	..	..	..	..	..
(*d*) As (*c*), with a priming coat of red-lead paint	1307	>5·0	>5·0	1490	4·6	>4·9	..	..	..	..	..	..	..	..	..
	1365	>5·0	>5·0	1460	4·6	>4·9	..	..	..	..	..	..	..	..	..
(*C*) OTHER COATINGS															
8. Tar/tallow/lime, applied hot ..	13	1·0	3·0	9	0·3	2·8	..	..	..	..	..	..	..	..	..
	11	1·0	2·5	14	0·3	2·8	..	..	..	..	..	..	..	..	..
9. Vitreous enamel	231	>5·0	>5·0	233	>4·9	>4·9	221	>5·2	>5·2	..	..	..	..	..	..
	287	>5·0	>5·0	229	>4·9	>4·9	239	>5·2	>5·2	..	..	..	..	..	..
10. Rubber/wax sheathing	346	4·6	>5·0	..	..	..	344	3·5	>5·2	..	..	..	..	..	..
	350	4·6	>5·0	..	..	..	348	3·5	>5·2	..	..	..	..	..	..

* Details of the coatings will be found in the Appendix.

Notes: *W* = weight of coating, grammes per specimen. In the case of paints the total weight of the dry film is given.
Lf = life of coating to grade "fair," years. Mean value for the front and back surfaces.
Lo = Life of coating to grade "failed," years. Mean value for the front and back surfaces.

It may be added that discoloration was more pronounced where red oxide paint alone had been used than where red-lead paint had been used as a primer, although this is not evident from the photographs.

(*b*) Over Weathered and Wire-Brushed Steel

The painting system in this case was the same as for group 1 (*e*) of Table III. This was a proprietary scheme comprising two coats of paints bound with special media in which synthetic resins were incorporated (*see* Appendix). The surface preparation of the specimens was different, however. They were exposed to the weather for four months until about 90% of the mill scale had been removed or loosened. After a thorough wire-brushing the surface was treated with a proprietary inhibitive wash of the phosphoric-acid type, which was supplied by the paint manufacturers; the priming and finishing coats of paint were then applied.

Bad results have been obtained from this process. Rusting commenced under the paint soon after exposure and spread rapidly; at all three home exposure stations the paint had reached the stage at which repainting would normally be required after about one year or less. As will be seen from Table III the average life to grade "failed" (5% rust) was 1·6 years at Sheffield, as compared with a life exceeding 5 years for the same painting scheme applied to a pickled surface. This behaviour is typical of painting schemes applied to weathered steel carrying residual mill scale. Thus the makers' claims that the use of this inhibitive wash would produce results on steel so prepared comparable with those obtained on steel descaled by pickling or shot-blasting have not been substantiated.

(*c*) Lanolin-Base Primers

These primers are lanolin-base paints hardened with synthetic resin; one is pigmented with zinc chromate, red oxide, and kaolin, the other with zinc chromate, zinc oxide, lead chromate, and kaolin. Two coats of primer were applied by brushing to specimens which had been pickled in dilute sulphuric acid with an inhibitor,* and additional specimens were prepared in which two coats of the Committee's standard red lead in linseed-oil primer were given. In all cases the painting was completed with two coats of the Committee's standard red oxide in linseed-oil finishing paint.

At Sheffield these specimens showed slight breakdown after 5 years, but at Llanwrtyd Wells and Congella signs of breakdown appeared after shorter periods of exposure. At Llanwrtyd Wells failure, once commenced, spread more rapidly on the specimens primed with the lanolin-base primers than on those primed with red lead; the former failed after an average period of exposure of 4·4 years, whereas the latter had not yet reached the point of failure at the time of the last inspection recorded here (4·8 years).

The performance of the painting schemes in question at Llanwrtyd Wells may be illustrated by the photographs, reproduced in Fig. 3, of three specimens primed with lanolin primer *J*, red lead in linseed oil, and lanolin primer *K*, respectively. The surface primed with red lead shows some rusting, but the lanolin primers *J* and *K* are both much worse in this respect.

On the whole, the results obtained with the lanolin-

* The specimens were of copper-bearing steel and the inhibitor used was di-*o*-tolylthiourea (*see* Appendix).

Table IV

RESULTS OF THE TOTAL IMMERSION TESTS IN SEA-WATER AT GOSPORT

Coating	Examination After Two Years' Immersion							
Process*	*W*†	Change in Weight,‡ g.	Grading§		Surface Area Rusted, %		Fouling‖	Notes
			Front	Back	Front	Back		
6. Sprayed Bitumen Mixtures:								
(a) On shot-blasted specimens	965	−559	*o*	*o*	95	80	*b w*	Mechanical damage may account for some of the breakdown of these coatings. Barnacle shells have penetrated the coating to a depth of 1 mm. or more, in some cases to the metal
	1105	+ 9	*f*	*f*	0	2	*B w*	
(b) On as-rolled specimens	1148	−928	*o*	*o*	95	90	*b w*	
	971	+ 10	*f*	*f*	2	2	*B w*	
(c) Bitumen + hard rubber on shot-blasted specimens	1061	−567	*f*	*o*	0	90	*B w*	
	908	+ 41	*f*	*f*	0	0	*B w*	
(d) Bitumen + hard rubber on as-rolled specimens	798	−165	*o*	*o*	5	20	*B w*	
	893	+ 11	*f*	*o*	2	5	*B w*	
(e) Bitumen + zinc on shot-blasted specimens	1538	− 82	*f*	*f*	1	2	*b w*	This coating is more adherent than the other bitumen coatings and less fouled by barnacles, which have penetrated to a depth of only about 0·4 mm.
	1493	+ 42	*x*	*f*	0	2	*b w*	
7. Sprayed Zinc (90%) + High-Silica-Glass (10%) on Shot-Blasted Specimens	133	− 28	*f*	*f*	0·3	0·1	*b*	About 2% of fine blisters and pits in the coating
	126	− 20	*f*	*f*	0·01	0·2	*b*	
9. Vitreous Enamel	240	− 0·3	*x*	*f*	0	0	*B*	About 20 small black spots in the coating on the back of each specimen
	234	+ 1	*x*	*f*	0	0	*B*	

* Details of the coatings will be found in the Appendix.
† Weight of coating, grammes per specimen.
‡ Change in weight on exposure, *i.e.*, the difference between the final weight, after scraping-off the marine growths and washing in fresh water, and the initial weight as exposed. A gain in weight is indicated by a positive figure, a loss by a negative figure. Pronounced losses in weight are due to loss of coating as a result of damage.
§ For grading scale see page 101.
‖ Hard-bodied fouling organisms remaining after cleaning are indicated thus:
Up to 50 barnacles per specimen, *b*; more than 50 barnacles per specimen, *B*;
up to 50 tube-worms per specimen, *w*; more than 50 tube-worms per specimen, *W*.

base primers are good, particularly in view of their much reduced weight as compared with red lead. It should be noted, however, that the general trend of the results of the Committee's experimental work during recent years is to show that red lead is particularly valuable as a priming paint when the steel has been prepared merely by weathering and wire brushing; when the surface of the steel is correctly prepared for painting, as in the present instance, the differences between red lead and other inhibitive, or even non-inhibitive, priming paints are less marked, and it may be found that other types of primer give better results.

(2) SPRAYED COATINGS

(a) Hot-Spray Paint

Specimens were coated by a special process in which a paint in a stand-oil/wax medium, containing no volatile constituent, is applied by the spray gun; the paint is warmed in order to give it the necessary fluidity. In this case the pigment consisted of a mixture of white lead, zinc oxide, and aluminium powder tinted with carbon black. The specimens were prepared for painting by pickling in hot dilute sulphuric acid containing an inhibitor.

This coating has behaved well at Sheffield, but at Llanwrtyd Wells and Calshot some staining has been observed, indicating that the steel is rusting under the paint film. Bearing in mind that only one coat of paint was applied, the results are good, although it is true that the weight of the paint film was high and comparable with that of the four-coat systems using lanolin-base primers and red oxide finishing coats referred to in 1 (*c*). The condition of the hot-spray paint after 5·6 years' exposure at Calshot is shown in Fig. 4, which should be compared with Fig. 2, representing the condition of ordinary air-drying painting schemes after this period.

(b) Cement/Asbestos

The cement/asbestos process virtually consists of enclosing the steel in a sheath of cement/asbestos, about $\frac{1}{4}$ in. thick, which is sprayed on as a wet slurry and covered for one day or more before exposure. A tack coat of bituminous emulsion is applied between the steel and the cement/asbestos. The surface preparation was varied within this group of specimens, so as to include all four possible combinations of as-rolled or pickled steel* with a red-lead priming coat or no priming coat.†

The nature of the coating renders it impossible to ascertain what is happening at the steel surface without stripping it, but, so far as could be judged from their external appearance, all the specimens were in good condition at the times of the last inspections (5 years), and no differences due to the variation in surface preparation beneath the cement/asbestos coating could be discerned. At the end of this period some slight surface erosion of the cement/asbestos coating itself had occurred, as can be seen from the photograph shown in Fig. 5.

* The sulphuric-acid/phosphoric-acid process was used.

† The bituminous tack coat was applied between the red lead and the cement/asbestos, after the red lead had been given ample time to dry.

(3) OTHER COATINGS

(a) Tar/Tallow/Lime Applied Hot

The tar/tallow/lime mixture used consisted of tar to which 1 lb. each of tallow and slaked lime had been added per gallon. It was brushed-on at about 200° C., to specimens carrying practically unrusted mill scale. This protective scheme has not behaved well either at Sheffield or Llanwrtyd Wells, failure having occurred within 3 years. This failure can be illustrated by the photograph, reproduced in Fig. 6, of part of the surface of a specimen to which the tar mixture had been applied, taken after 5·6 years' exposure at Llanwrtyd Wells. This photograph is comparable with the photographs of painting schemes proper shown in Figs. 1 and 3. If, however, the tar/tallow/lime mixture is regarded as a substitute for a single coat of paint, the results may be taken as satisfactory for a coating applied to steel in this surface condition.

(b) Vitreous Enamel

The vitreous-enamel coatings were applied to specimens of ingot iron of the same size as the others ($15 \times 10 \times \frac{3}{8}$ in.), because difficulties were experienced in securing the adhesion of ordinary vitreous enamels to the standard mild steel, containing 0·2% of carbon, used for the remainder of the tests. Three coats of enamel were applied, following the usual industrial procedure.

As was anticipated, the vitreous enamel has proved markedly resistant to atmospheric attack, and, apart from slight mechanical damage which has chipped the coating on some of them, all the specimens concerned were in perfect condition after 5 years' exposure. It is clear that vitreous enamel should prove an excellent protective coating for articles that can be coated in this way, involving heating to 800–900° C. and which are not exposed in service to risk of mechanical damage. There would, therefore, appear to be scope for investigations directed towards increasing the range of applicability of vitreous enamelling to the protection of ferrous metals in general. In this connection three possibilities spring to mind immediately, namely:

(i) The application of vitreous enamel, without heating the article coated, by spraying, using a modified form of metal-spraying pistol.

(ii) The development of non-brittle enamels, less subject to injury by mechanical damage.

(iii) Research on rendering vitreous enamels less sensitive to the effects of variation in the basis material, *i.e.*, the ideal would be to render it equally easy to apply vitreous enamel over cast iron, mild steel, wrought iron, or ingot iron, without regard to differences in the composition of the basis ferrous material.

(c) Rubber/Wax Sheathing

Rubber/wax sheathing is another "enveloping" process, in which the article to be protected is covered with a sheath of material prepared from raw rubber, waxes, and resins. The material is available as sheets or tubes, but for the purpose of the present tests the specimens were coated with $\frac{1}{16}$-in. thick sheets

Red oxide in linseed oil, 2 coats | Red lead in linseed oil, 1 coat ; red oxide in linseed oil, 3 coats | Proprietary paint, 2 coats

Fig. 1—Condition of two-coat painting schemes applied to pickled mild steel after 5·6 years' exposure at Llanwrtyd Wells. All specimens pickled by the duplex sulphuric-acid/phosphoric-acid process before painting. Although not evident from the photographs, the left-hand specimen, painted with red oxide, was much more discoloured than the middle specimen, painted with red lead followed by red oxide

Red oxide in linseed oil, 2 coats | Red lead in linseed oil, 1 coat ; red oxide in linseed oil, 3 coats | Red lead in linseed oil, 2 coats ; red oxide in linseed oil, 3 coats

Fig. 2—Condition of painting schemes applied to pickled mild steel after 5·6 years' exposure at Calshot. All specimens pickled by the duplex sulphuric-acid/phosphoric-acid process before painting

Lanolin-base primer J, 2 coats ; red oxide in linseed oil, 2 coats

Red lead in linseed-oil primer, 2 coats; red oxide in linseed oil, 2 coats

Lanolin-base primer K, 2 coats ; red oxide in linseed oil, 2 coats

Fig. 3—Condition of painting schemes applied to pickled copper-bearing mild steel after 5·6 years' exposure at Llanwrtyd Wells. All specimens pickled in sulphuric acid before painting

Fig. 4—Condition of hot-spray paint coating applied to pickled mild steel after 5·6 years' exposure at Calshot. The specimen was pickled in sulphuric acid before the coating was applied

Fig. 5—Condition of sprayed cement/asbestos coating applied to mild steel after 5·6 years' exposure at Llanwrtyd Wells

Fig. 6—Condition of tar/tallow/lime coating applied hot to as-rolled mild steel after 5·6 years' exposure at Llanwrtyd Wells

applied by a combination of heat and pressure, special attention being paid to the joints which ran along the edges. The coating was applied directly over the mill scale on the steel.

At Sheffield the specimens were still intact after 5 years, the only visible changes being slight surface cracks in the coating and a change in colour from black to brown. After the same period at Calshot the coating had cracked and commenced to peel from the edges.

It would seem that the chief uses of this coating material for protection against atmospheric corrosion will be found in cases where it is necessary to mould a flexible insulating coating *in situ* to cover an article of special shape.

DISCUSSION OF THE RESULTS OF THE TOTAL-IMMERSION TESTS IN SEA-WATER

(1) FOULING

When the specimens at Gosport were removed after 2 years' immersion in sea-water they were covered with a thick layer of mud, slime, sea-squirts, and sponges, in which barnacles and tube-worms were hidden. Marine plants were absent, presumably because the specimens had been shielded from daylight by the landing-stages under which they were exposed. The bitumen + zinc coated specimens were less fouled than the other bituminous coatings, while the specimens sprayed with zinc + high-silica-glass were only lightly fouled. This is illustrated by the approximate indications given in Table IV of the number of barnacles and tube-worms that remained attached to the specimens after cleaning ; it will also be noted that the vitreous-enamelled specimens were badly fouled by barnacles, but not at all by tube-worms.

These observations are examples of the known toxicity of zinc and zinc coatings to fouling organisms, although it should be emphasized that this toxicity is inadequate to permit the use of zinc as an effective anti-fouling agent.

(2) PROTECTIVE SCHEMES

The relative merits of the protective schemes tested are best revealed in Table IV, by the gradings of the surfaces, together with the figures for rusting ; the changes in weight are of limited significance, and no pitting of the steel base was observed.

(*a*) Sprayed Bitumen Mixtures

The behaviour of the sprayed bitumen coatings proved erratic. Only one surface was adjudged to be in good condition after 2 years' immersion ; the remainder were graded as "fair" or "failed," in approximately equal proportions. The breakdown observed may have been initiated in part by damage during transport and handling, but some of the coatings were blistered, and the steel surface was found to be rusting even in places where the coating was apparently intact. The surface of these coatings was rough and seemed to encourage the settlement of fouling organisms ; it was noticed that the barnacle shells had penetrated into the coating, in some cases right down to the metal. Within the rather wide range of variation of duplicate results, such variations in surface preparation and in the composition of the sprayed bitumen as were tested cannot be said to have had an appreciable effect, except for the influence of the admixture of zinc in reducing fouling, to which reference has already been made ; the bitumen and zinc mixture seemed also to be more adherent than the other bitumen coatings. A more extensive investigation under conditions in which the possibility of mechanical damage was excluded would be necessary to assess the relative merits of these and other bitumen coatings for the protection of steel in sea-water.

(*b*) Sprayed-Zinc/High-Silica-Glass

The sprayed coatings of powdered zinc + high-silica-glass were in fair condition, with an average surface area rusted of about 0·15%, and had not yet failed. Their performance did not differ appreciably from that of sprayed coatings containing zinc only, exposed at the same time,[1] (*loc. cit.*, pp. 241P–242P), so that the addition of the glass has proved to have had no effect.

(*c*) Vitreous Enamel

The performance of vitreous enamel was outstanding, and at the end of the exposure period the specimens concerned were practically intact.

SUMMARY

An account is given of the field tests of the Protective Coatings (Corrosion) Sub-Committee on protective schemes consisting of paints or other non-metallic coatings for ordinary irons and steels. These tests are being carried out in parallel with tests on metallic coatings, an account of which has already been published.[1] The protective schemes were applied to 15 × 10 × ⅜-in. flats of mild steel, or, in one case, of ingot iron, and the coated specimens were exposed both to the atmosphere and to total immersion in sea-water. The following interim conclusions are drawn from the results of observations over periods of up to 5 years in the case of the atmospheric tests, or of 2 years in the case of the sea-water tests :

(1) Two-coat painting schemes applied to specimens pickled by the sulphuric-acid/phosphoric-acid process, which involves a final dip in 2% phosphoric acid, were in good condition after 5 years in the atmosphere.

(2) In contrast to this, bad results were obtained when one of these painting schemes was applied to specimens that had been partially descaled by weathering and then wire-brushed ; in this case failure occurred in 1·6 years or less. Treatment of this type of surface with a proprietary inhibitive wash containing phosphoric and chromic acids before painting failed to improve matters.

(3) Lanolin-base paints hardened with synthetic resin and containing chromate pigments have given good results as primers over pickled steel in atmospheric tests.

(4) A special process in which a paint containing no volatile matter is sprayed-on hot has given good results at Sheffield but shows signs of breakdown after 5 years in non-industrial atmospheres.

(5) A tar/tallow/lime mixture, applied hot, failed to protect the steel for more than 3 years in an industrial atmosphere, but this life is comparable with that which would have been obtained from a single paint coat of approximately the same weight.

(6) In the case of coatings of sprayed cement/asbestos, $\frac{1}{4}$ in. thick, no external sign of corrosion of the encased steel was visible after 5 years.

(7) Vitreous-enamel coatings were still in perfect condition, apart from minor mechanical damage, after 5 years in industrial or non-industrial atmospheres, or 2 years in sea-water.

(8) A rubber/wax sheathing has given good results in an industrial atmosphere, but has not behaved so well in a marine atmosphere.

(9) Heavy coatings of sprayed bitumen, alone or mixed with rubber or zinc, gave fairly good results when continuously immersed in sea-water.

(10) The addition of glass to zinc coatings sprayed by the powder pistol had no appreciable effect upon their performance when continuously immersed in sea-water.

Appendix

PREPARATION OF THE SPECIMENS

MILD-STEEL FLATS

The specimens were each 15 × 10 × $\frac{3}{8}$ in., being sawn from 10 × $\frac{3}{8}$-in. rolled flats of basic-open-hearth mild steel of the following analysis :

Carbon	0·2%
Manganese	0·6%
Silicon	0·05%
Phosphorus	0·04%
Sulphur	0·03%
Copper	0·03%
Nickel	0·04%

Full details of the manufacture of the steel and of the rolling and cutting-up of the specimens were recorded, and have been summarized in an earlier paper.[1] Representative specimens were subjected to a chemical and metallurgical examination, which showed that the material was normal and typical of mild steel of this class.

APPLICATION OF THE COATINGS

With the exception of the painting schemes, which were applied in the laboratory, all the protective coatings were applied at works where the processes were in commercial use. In cases where a preliminary treatment, such as pickling or shot-blasting, was given, the specimens were weighed in the works after this treatment. The weights of the coated specimens were determined in the laboratory, and the coating weights were deduced by difference. The values so obtained are considered to be accurate to about 0·5 g.

Through the courtesy of the firms concerned, each set of specimens was prepared in the presence of one of the authors, who recorded details of the methods used for the application of the coatings. The following abbreviated accounts of the various processes are based on these records, and in each case are published with the consent of the companies who kindly provided the necessary facilities.

Table V

FORMULATIONS OF THE STANDARD RED LEAD AND RED OXIDE IN LINSEED-OIL PAINTS

Ingredient	Red Lead, wt.-%	Red Oxide, wt.-%
Red lead (Pb_3O_4 95% min.) ...	81·5	...
Iron oxide (natural Spanish) ...	...	62·2
Refined linseed oil	8·1	21·1
Boiled linseed oil (Pb 0·15%, Mn 0·05%)	8·1	12·4
White spirit	2·4	1·9
Lead–manganese terebine ...	...	2·5

(1) PAINTING SCHEMES

All the coatings classed as paints in Table I were brush-applied in the laboratory under good conditions, ample drying time being allowed between successive coats and between the final coat and exposure. The weights of wet paint applied to the specimens were determined by weighing the paint kettle and brushes before and after painting each specimen ; the weights of dry paint were determined by weighing the specimen before painting and again after each coat of paint had dried.

(*a*) Over Steel Pickled by the Sulphuric-Acid/Phosphoric-Acid Process

The specimens pickled by the sulphuric-acid/phosphoric-acid process were prepared for painting at a works where the duplex sulphuric-acid/phosphoric-acid process* is in operation on a large scale. They were descaled by immersion for 15–20 min. in dilute (5–10%) sulphuric acid at 60–65° C., rinsed by dipping twice into water at 60–65° C., and finally immersed for 3–5 min. in a bath containing approximately 2% of free phosphoric acid and 0·3–0·5% of iron, heated to 85–90° C. On removal from the final bath the specimens dried rapidly, leaving a dull-grey film of iron phosphate on the surface. This film by itself

* *See* H. B. Footner, *The Iron and Steel Institute*, 1938, *Special Report No. 21.*

Table VI

FORMULATIONS OF THE LANOLIN-BASE PRIMERS

Ingredient	Specification	Primer *A*. wt.-%	Primer *B*. wt.-%
Lanolin	D.T.D. 122*B*	20·0	20·0
Modified phenolic resin	D.T.D. 375*A**	13·3	13·3
Zinc chrome ...	D.T.D. 377	16·7	16·7
Lead chromate ...	B.S.S. 282	...	7·5
Red oxide of iron	B.S.S. 2*D*.28	10·0	...
Zinc oxide... ...	B.S.S. 2*D*.27	...	7·5
Kaolin	D.T.D. 374	6·6	6·6
White spirit ...	B.S.S. 244	16·7	14·2
Solvent naphtha...	B.S.S. 479*A*	16·7	14·2

* The actual resin used was Albertol 111*L*.

has only slight protective value, but forms an excellent basis for paint.*

The pickled specimens were painted in the laboratory in five different ways, as detailed below. The approximate total thickness of the dry paint film is given in a few cases, the figures being an average for all the specimens painted by each process :

(*a*) Two coats of red oxide in linseed-oil paint. Approximate film thickness 2·1 mils.

(*b*) One coat of red lead in linseed-oil paint, followed by three coats of red oxide in linseed-oil paint.† Approximate film thickness 3·7 mils.

(*c*) Two coats of red lead in linseed-oil paint, followed by three coats of red oxide in linseed-oil paint.† Approximate thickness 4·6 mils.

(*d*) Two coats of proprietary synthetic paints. The priming paint contained white lead, basic lead chromate, red oxide, red lead, and magnesium silicate in an alkyd varnish, and the finishing paint black oxide of iron and carbon black in an alkyd varnish.‡

(*e*) A special proprietary process used in the United States. This comprised a preliminary surface treatment with an inhibitive wash of the phosphoric-acid/chromic-acid type, followed by the application of two coats of paint in glyptal/phenolic/tung-oil varnish, the priming paint being pigmented with equal parts of zinc chromate and red oxide, and the finishing paint with aluminium.‡

The red-lead and red oxide paints used in painting schemes (*a*), (*b*), and (*c*) were the standard paints used in the Committee's earlier painting tests ; their formulations are given in Table V.

(*b*) Over Weathered and Wire-Brushed Steel

A test over weathered and wire-brushed steel was conducted to check the claim made by the proprietors of the American process referred to in the previous sub-section, that their process was efficacious on weathered steel. For this purpose specimens were exposed in the as-rolled condition to atmospheric weathering in Birmingham for 4 months. They were then thoroughly wire-brushed by hand, after which approximately 10% of the area was found to be covered with firmly adherent mill scale. Finally, the protective process comprising an inhibitive wash and two coats of paint was applied.

(*c*) Lanolin-Base Primers

In order to compare priming paints of the hardened-lanolin type with the more widely known red lead in linseed oil, tests were made on the following painting schemes :

(i)	Lanolin primer *A*	2 coats.
	Red oxide in linseed oil ...	2 coats.
(ii)	Lanolin primer *B*	2 coats.
	Red oxide in linseed oil ...	2 coats.
(iii)	Red lead in linseed oil ...	2 coats.
	Red oxide in linseed oil ...	2 coats.

Paints of this type had been developed at the Royal Aircraft Establishment for other purposes, and the samples used in the present tests were kindly supplied by the Establishment. Details of their formulation are given in Table VI.

Owing to shortage of material these tests were not made on specimens of the mild steel used for the bulk of the present tests but on specimens of copper-bearing steel, containing 0·2 or 0·5% of copper, as used in earlier tests of the Committee. Before painting, the mill scale was removed from the specimens by pickling them in warm dilute sulphuric acid.

(2) SPRAYED COATINGS

All the sprayed coatings referred to below were applied under works conditions.

(*a*) Hot-Spray Paint

In the hot-spray paint process§ a special paint is used, which contains no volatile constituent and is solid at ordinary temperatures. A moulded block of the paint is placed in the container of a specially designed spray pistol, where it is melted electrically.

* If the process is to yield the best results, the pickled steel should be given a coat of priming paint immediately it is dry, whilst still warm. In the case of the specimens some delay occurred before painting, but they were carefully stored during the interval.

† It was originally intended to apply one coat of red oxide in linseed-oil paint in case (*b*), and two coats in case (*c*). On drying, the finishing paint, which had been applied thinly, was found to have cracked over the red-lead primer and, to obtain a satisfactory finish, it proved necessary to apply further coats of paint, up to a total of three coats in each case ; 9% by weight of linseed oil was added to the last of these coats. The total thickness of the three coats of red oxide in linseed-oil paint applied to these specimens was approximately 20% greater than that of the two coats applied in case (*a*).

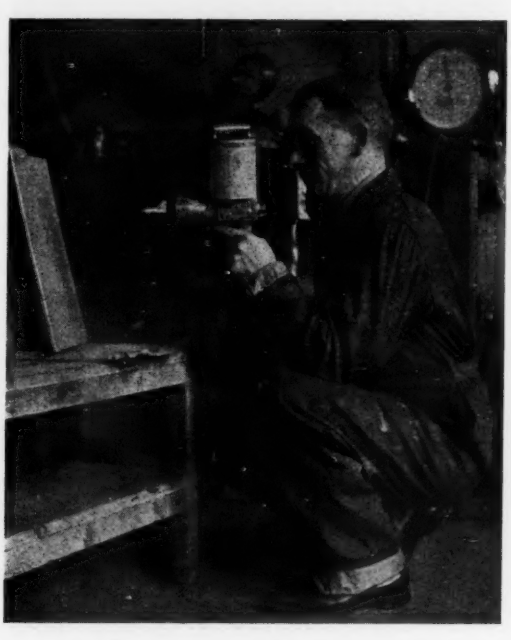

Fig. 7—Application of hot-spray paint

‡ In the case of these proprietary painting schemes the data needed to calculate the paint-film thicknesses are not available.

§ British Patents Nos. 275,477 and 386,101.

The molten paint flows through an electrically heated pipe to a central nozzle, surrounded by an annular air-blast, which atomizes and sprays the paint, as shown in Fig. 7. The paint and air nozzles are surrounded by a number of small gas-flames. These flames dry and warm the surface being sprayed, and also initiate polymerization-drying of the paint. The pistol uses electrical power, compressed air, and a fuel gas, such as butane, and it is claimed that 6–8 sq. yd./hr. can be sprayed.

The specimens were prepared by pickling in warm dilute sulphuric acid containing an inhibitor.* A single coat of paint was applied to them as described above ; this had the following composition :

	Wt.-%
White lead	25
Zinc oxide	9
Aluminium powder	5
Carbon black	1
Stand oil	40
Montan wax	13
Modified phenolic resin† ...	7

† The actual resin used was Albertol 111*L*.

The paint film is dry as soon as it is sprayed, but requires a few days to become hard. Minor damage to the specimens, caused during handling and transport, was repaired by spreading warmed Montan wax over the affected areas.

(*b*) Cement/Asbestos

Cement/asbestos coatings are sometimes applied to steel, chiefly for heat- and sound-insulation purposes. It has been found, however, that the protection against corrosion afforded by these coatings is good, and they are now being used for this purpose as well. For the present tests, coatings of this type were applied to specimens in the following surface conditions :

(i) Pickled by the duplex sulphuric-acid/phosphoric-acid process.

(ii) As (i), followed by painting with a single coat of red lead in linseed-oil paint.

(iii) As-rolled, *i.e.*, with the intact mill scale.

(iv) As (iii), but painted with a single coat of red lead in linseed-oil paint.

In cases (ii) and (iv) the red-lead paint was allowed ample time to dry hard. Immediately before spraying-on the asbestos/cement, each specimen was coated with an aqueous bitumen emulsion (60% of petroleum + 40% of water, stabilized with bentonite). A mixture of asbestos ($\frac{1}{2}$-in. fibre) and cement slurry (23% of Portland cement and 77% of water), in equal parts by weight, was then sprayed on as shown in Fig. 8, to a depth of $\frac{1}{4}$ in. The shredded asbestos and the cement slurry were sprayed through separate convergent nozzles by air blasts, the two streams mixing at a distance of 9 in. from the nozzles.‡ The twin nozzles were held 10–12 in. from the work, so that the cement slurry and the asbestos fibre were mixed immediately before striking the surface. After spraying, the coating was pressed and trowelled smooth whilst still plastic, then left for a week to dry and harden.

The dry coating had a fairly smooth white-grey surface and the specimens were, in effect, completely sheathed in the mixture to a thickness of $\frac{1}{4}$ in.

(*c*) Bitumen Mixtures

Spraying processes as applied to bitumen have the value that they render it possible to produce a continuous bitumen coating on any size or shape of article ; for instance, coatings of this type are often applied to dock-gates, piles, and piers. Accordingly, three coatings of sprayed bitumen were included in the present tests, namely bitumen alone, bitumen mixed with hard rubber, and bitumen mixed with zinc. These were applied by means of a modified form (a larger model) of the powder pistol used for metal spraying.

Some of the specimens were prepared for spraying by shot-blasting, others received no special treatment and were coated over the as-rolled surface with the mill scale. In all cases a thin tack-coat of bitumen was first applied by brushing, using a solution in naphtha or an emulsion in water ; the various bitumen mixtures were then sprayed on in the same way as metal powders. The coating thickness on the flats was approximately $\frac{1}{16}$ in. for bitumen alone or bitumen mixed with rubber, but $\frac{1}{8}$ in. for bitumen + zinc ; in all cases the edges were given an extra-thick coating.

Fig. 8—Application of sprayed cement/asbestos

* The inhibitor used was di-*o*-tolylthiourea, 0·05 wt.-% of which was added to the acid solution. The material required was supplied through the kindness of the Imperial Chemical Industries, Ltd.

‡Alternatively, a mixture of asbestos and cement in the required proportions can be sprayed on with water only, through a single nozzle.

After spraying, the coating was flamed over, to weld together the sides and edges and to smooth the surface as far as possible. The spraying rate on the flats was approximately 1 sq. ft./min.* The sprayed coatings had a coarse granular glossy-black surface.

The powders used were graded 30–200-mesh, and had the following compositions :

(i) *Bitumen :*		
Oxidized bitumen compound	...	40%
Silica filler	...	60%
(ii) *Bitumen + Rubber :*		
Oxidized bitumen compound	...	45%
Silica filler	...	45%
Hard rubber	...	10%
(iii) *Bitumen + Zinc*		
Oxidized bitumen compound	...	40%
Silica filler	...	35%
Zinc powder	...	25%

In view of the fact that coatings of this type would be mainly used for the protection of steel under conditions of complete immersion, the specimens concerned were tested only in sea-water, and no atmospheric exposures were made.

(*d*) Zinc + High-Silica Glass

It had been suggested that the addition of high-silica glass to sprayed zinc coatings applied by the powder-pistol process would be of value when such coatings were used for the protection of steel immersed in sea-water. A number of specimens coated in this way were therefore prepared for test. The steel was first descaled by shot-blasting, then sprayed with a powder containing 90% of zinc mixed with 10% of high-silica glass, using the standard powder pistol as used for metal spraying. The nominal coating thickness was 5 mils. The coating had a smooth mat light-grey metallic appearance, similar to that of zinc-sprayed coatings.

* It is claimed that, using the largest model of the pistol, bitumen can be sprayed to a thickness of $\frac{1}{16}$ in. in one pass at rates of up to about 60 sq. ft./hr., depending on the shape and size of the work.

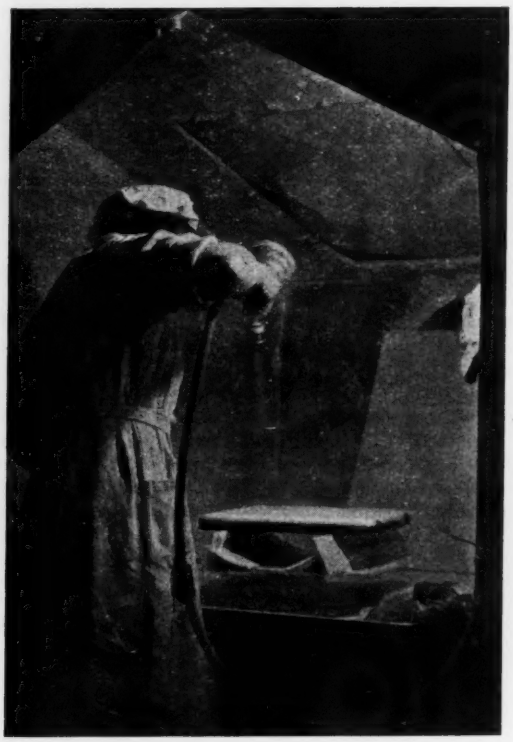

Fig. 9—Spraying vitreous enamel

Fig. 10—Specimens coated with vitreous enamel leaving the furnace

(3) OTHER COATINGS

Except for the tar/tallow/lime, which was brushed on in the laboratory, the coatings described below were applied under industrial conditions.

(*a*) Tar/Tallow/Lime Applied Hot

Mixtures based on tar are widely used for painting steelwork in locations where appearance is not a major consideration, and also for steelwork immersed in sea-water. Four as-rolled specimens were coated with the following mixture, which is a formulation used by the Great Western Railway Company :

Dehydrated coal tar (horizontal retort)	1 gal.
Tallow	1 lb.
Slaked-lime powder	1 lb.

The mixture was heated until it boiled at 190–200° C., then brushed on, using an old stiff brush.

(*b*) Vitreous Enamel

Vitreous-enamel coatings are widely used for decorative and protective purposes on domestic utensils, stoves, water-heaters, chemical and dairy plant, road signs, and to some extent on the exteriors of buildings.

It was found that ordinary vitreous enamel could not be applied successfully by the usual methods to the 15 × 10 × $\frac{3}{8}$-in. specimens of the mild steel prepared for this investigation ; steel of this composition is not normally used in the form of such a thick section for enamelling. Accordingly, specimens of ingot iron of the same size, 15 × 10 × $\frac{3}{8}$ in.,† were used for the preparation of the vitreous-enamelled

† The specimens were taken from the Committee's stock of ingot-iron specimens. The preparation of this material is described in the Second Report of the Corrosion Committee, *The Iron and Steel Institute*, 1934, *Special Report No. 5*, pp. 33 and 124.

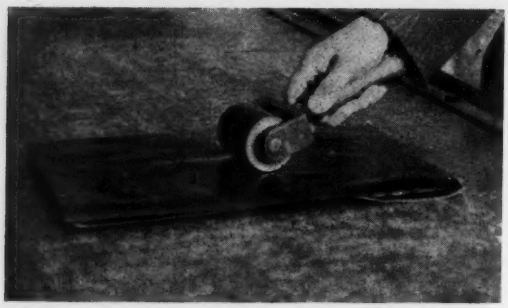

Fig. 11—Application of rubber/wax sheathing

specimens ; the analysis of this material is as follows :

Carbon	0·03%
Manganese	0·05%
Silicon	0·03%
Phosphorus	0·01%
Sulphur	0·04%
Copper	0·04%

The ingot-iron specimens were first heated in a furnace at 840° C. for 2½ min. to degas them and burn off any oil or grease on the surface. After cooling, they were pickled in hot dilute hydrochloric acid, rinsed twice in water, once in hot soda-ash solution, and dried by warm air. Three coats of enamel were then applied and fused separately. The enamel was applied to the specimens as a suspension in water stabilized by clay, known as a " slip." The first or " grip " coat, grey in colour, was applied by swilling the specimens in a bath of the suspension. The specimens were then dried in an oven, leaving a coating of the unfused powdered enamel. Damaged spots were touched-up with a small brush, after which the specimens were placed in a furnace at 880–890° C., to fuse the enamel. The specimens had to be left in the furnace for 15–20 min. ; thin sheets attain the furnace temperature much more quickly, and enamel on them becomes fused in a few minutes.

The second coat, white in colour, was applied by spraying the suspension on to the specimens, as shown in Fig. 9. After drying, the specimens were fired at 790° C. for 13 min. A third coat followed, the same white enamel being used again ; this was sprayed on, dried in the same way, and fired for 12½ min. at 780° C. The specimens on leaving the furnace are shown in Fig. 10.

When completed, the specimens had a smooth, white, glassy surface. Details of the composition of the actual vitreous enamels applied to the specimens are not available, but typical analyses of a ground or grip-coat enamel and of a white acid-resisting finishing enamel, which may be regarded as representative of the types of material used, are given in Table VII. It will be noted that the grip coat contains cobalt oxide, this being necessary to secure good adhesion to the steel base ; the manganese and nickel oxides present also help in this way. The white enamel contains titanium and antimony oxides as opacifiers.

(c) Rubber/Wax Sheathing

The material used for the rubber/wax coating is a mixture of raw rubber and natural waxes and resins.

Table VII

TYPICAL ANALYSES OF VITREOUS ENAMELS

Constituent	Ground Coat, wt.-%	Acid-Resisting White Cover Coat, wt.-%
SiO_2	51·3	50·7
Al_2O_3	6·8	2·0
B_2O_3	13·0	11·8
F (as fluoride) ...	3·4	1·2
Na_2O	16·2	14·9
K_2O	3·1	1·0
CaO	5·0	2·6
Co_2O_3	0·5	...
NiO	0·6	...
MnO_2	0·6	...
Sb_2O_3	...	7·7
TiO_2	...	8·4
	100·5	100·3

It is available as sheets and tubes, which are applied to metal or other surfaces either by the use of rubber adhesives or by heat and pressure. It is largely used for protecting electrical equipment and chemical apparatus.

The specimens were coated in the as-rolled condition, over the mill scale. The metal was first warmed to about 240° F. by radiant heat, the temperature being measured by a contact thermocouple. A $\frac{1}{16}$-in. thick sheet of the material was then rolled into contact with each face of the specimen, allowing an ample margin all round, as shown in Fig. 11. The sheets were trimmed and the edges bent around the edges of the specimens, the sheets from the two sides overlapping to some extent. The two thicknesses of material on the edges of the specimen were then softened and jointed by ironing with a bent steel tool warmed to 320–330° F. This method of jointing pieces of the material by ironing is a valuable feature of the process, and by its use the specimens were completely sheathed in the rubber/wax mixture.* They then had the smooth black surface with crocodile markings typical of the coating sheet material.

LABORATORY EXAMINATION OF THE COATED SPECIMENS

In addition to the determinations of the weights of the coatings, made by weighing the specimens at appropriate stages, all the coated specimens were inspected visually in the laboratory before exposure. This examination included a record of the appearance of the coating as regards colour, gloss, roughness, and surface markings, and a note of the nature and position of any defects.

References

1. J. C. HUDSON and T. A. BANFIELD : *Journal of The Iron and Steel Institute*, 1946, No. II, pp. 229P–264P.
2. FIRST REPORT OF THE MARINE CORROSION SUB-COMMITTEE : *Journal of The Iron and Steel Institute*, 1943, No. I, pp. 399P–401P.

* The material can also be wiped-down at the edges, by a hot iron, so that moisture cannot seep between the metal and the sheet.

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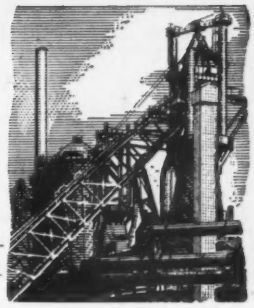

Gaseous and Liquid Fuels in Iron and Steel Works

ENGINEERING ASPECTS OF DISTRIBUTION AND UTILIZATION

By J. B. R. Brooke, A.M.I.C.E., and J. S. Bryan, A.M.I.C.E.

THE engineering aspects of the distribution and utilization of gaseous and liquid fuels in iron and steel works comprise a subject which can be approached either from the point of view of the manufacturer of equipment or from the point of view of the requirements which result from experience gained on existing steelworks. In effect the present paper describes a North Lincolnshire steelworks which commenced operations 35 years ago and which has been altered and enlarged ever since, and the paper is intended to be complementary to contributions which it is hoped representatives of plant manufacturers will make.

DISTRIBUTION OF BLAST-FURNACE GAS

The distribution of fuels throughout the works at current rates of operation is shown in Fig. 1. The cleaning system employed is of the wet type, the gas being cleaned in stages in accordance with the degree of cleanliness considered necessary at the time of installation. Obviously, from a maintenance engineer's viewpoint, the nearer the cleaning plant is to the blast-furnaces, in order to restrict the crude-gas system, and the more efficient the cleaning, the better he will like it.

A description is given of the existing system and, wherever possible, the problems met with in distribution and the means employed in overcoming these difficulties.

Under the rebuilding programme the wet-washing system is being replaced by an electrostatic filter plant, and therefore a major portion of the problems encountered at the present time will not apply. A brief description of the intended distribution is included.

EXISTING WET-CLEANING SYSTEM

The system is split into two stages, the first at each furnace comprising dustcatchers, water towers, fans with water sprays, and separators. The gas immediately after the separators has a dust content of from 0·35 to 0·45 g./cu. m., a temperature of 113° F., and a pressure of 14 in. W.G. From the separators the gas is delivered by mains at each furnace into a common main and is metered in each cross-main by a Kent ring-balance-type meter working from the differential across an orifice plate. Connections are taken off the underside of the common main for each stove and are butterfly controlled.

Owing to deposition in the mains, the gas at the boilers and the suction side of the final-stage cleaning plant has a dust content of 0·25 to 0·35 g./cu. m. It will be noted that even allowing the minimum deposition in the main of 0·1 g./cu. m., this quantity is equivalent to 0·3 cwt./hr., and therefore extreme care has to be taken in the cleaning of mains.

The gas in this condition is burnt at the boilers. The main continues to the gas-holder, which is connected in parallel, and is extended to the soakers. The dust content at the holder inlet is 0·25 g./cu. m., which raises problems detailed later in this paper.

The final stage of cleaning consists of Cockerill and Thiesen washers feeding the coke ovens and powerhouse, respectively. There are two Cockerill washers, which are run alternately, each capable of cleaning 1,250,000 cu. ft./hr. at a pressure of 12 in. W.G. The gas to the coke ovens must not contain more than 0·02 g. of dust per cubic metre, and usually contains 0·014 g./cu. m. Extreme care is taken with this plant, and regular dust tests are taken on the delivery leg to ensure that the specified conditions are being attained. The coke-oven leg is fitted with an Askania low-pressure regulator which maintains constant gas

Manuscript received 3rd December, 1947.

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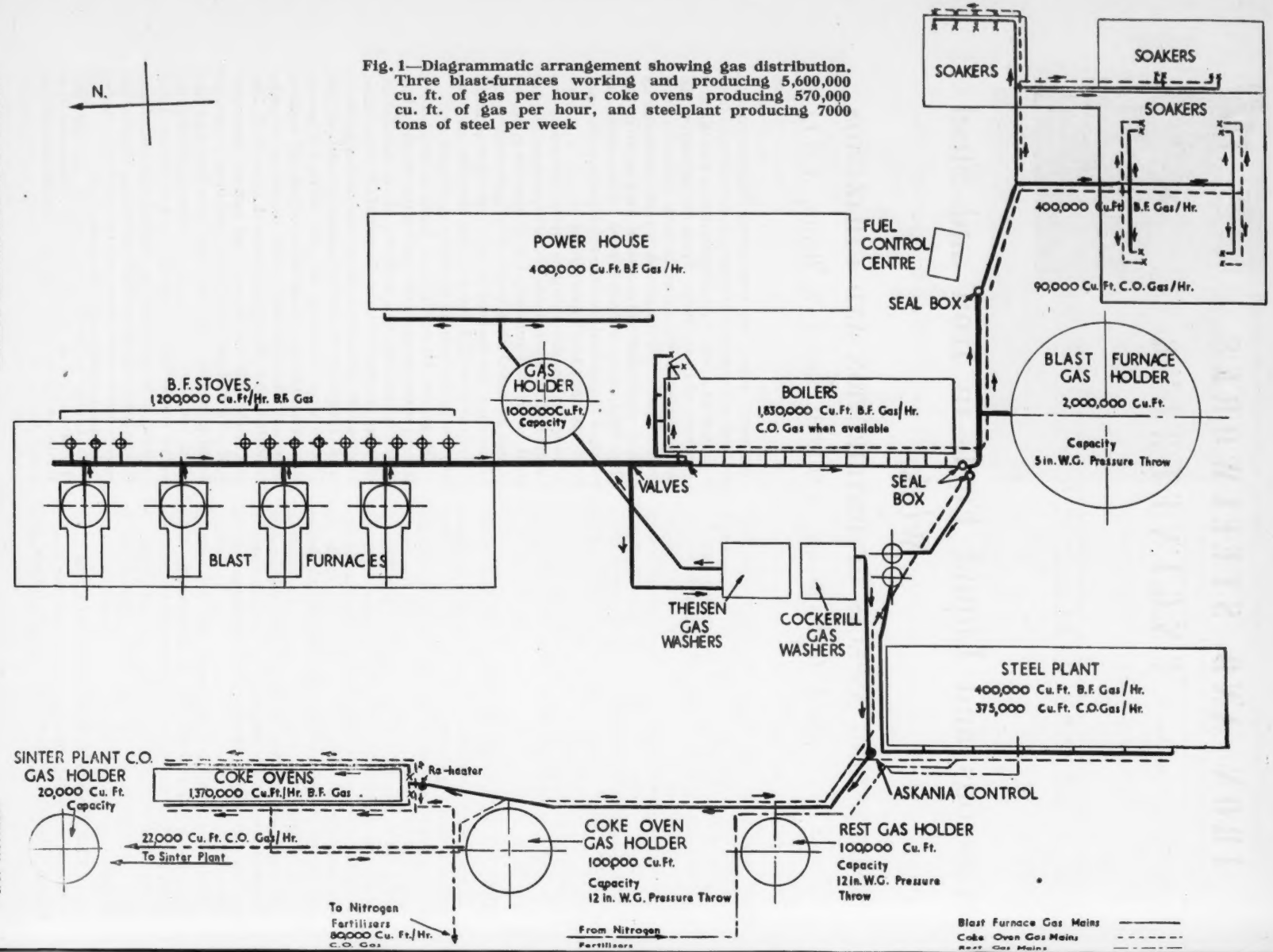


Fig. 1—Diagrammatic arrangement showing gas distribution. Three blast-furnaces working and producing 5,600,000 cu. ft. of gas per hour, coke ovens producing 570,000 cu. ft. of gas per hour, and steelplant producing 7000 tons of steel per week

pressure for distribution to each side of the battery. The Thiesen washers clean the gas down to 0·02 g. of dust per cubic metre, for use on the gas engines in the power-house, and deliver into a clean-gas-holder before distribution. This clean-gas-holder is a single-lift column-guided type of 100,000-cu. ft. capacity, and throws a pressure of 4½ in. W.G. when full.

The steelplant feed is first cooled over vertical towers and controlled by an Askania regulator to a pressure of 4 in. W.G.

Approximate water requirements per 1000 cu. ft. of gas for the different stages are as follows:

Tower washers	7 gal.
Fans	5 gal.
Cockerill washers	6 gal.
Thiesen washers	5·5 gal.

These quantities entail with the present gas makes a quantity of 130,000 gal. of water per hour in circulation.

DRY-CLEANING SYSTEM

For conversion to dry cleaning the system shown in Fig. 1 will be modified by locating the dustcatchers at the west of the stoves and the electrostatic filter plant north of the power-house. The main from each final dustcatcher will connect into a common main discharging into the primary washers. Isolating valves of the thermal-expansion goggle type are being installed after each dustcatcher. The velocities in the crude-gas system will be maintained at 25–30 ft./sec., and adequate provision will be made for cleaning the mains.

The delivery leg from the cleaning plant will be connected to the existing 2,000,000-cu. ft. gas-holder, and supplies to the coke ovens, stoves, power-house, steelplant, and boilers will be taken off this leg. The soakers will be supplied as previously, by an extension main from the gas-holder. With these modifications in layout all users will be metered, and details of the consumption will be transmitted to the central control room.

BLAST-FURNACE-GAS-HOLDER

The blast-furnace-gas-holder is a waterless M.A.N. type, of 2,000,000-cu. ft. capacity, connected in parallel with the gas main and situated as near as possible to the centre of the works system. It is designed to throw a pressure of 5 in. W.G., but the piston can be weighted to throw up to 12 in. W.G. pressure. Therefore, in addition to providing gas storage, covering fluctuations in gas consumption and production, the gas-holder has a stabilizing effect on the gas pressure. The possibility of the piston being brought down to the bottom and collapsing owing to the suction of the works system was guarded against by suspending a rectangular plate by chains to the piston and 6 ft. 0 in. down from it, so that when the piston is 6 ft. 0 in. from the bottom this plate seals-off the outlet orifice.

The size of the holder was arrived at on the basis that its capacity should have some relation to the gas production of the largest blast-furnace, which in 1930 was normally about 1,800,000 cu. ft./hr. Hence with a furnace off for tuyering or any other purpose, the holder would take the place of it for a period dependent, obviously, on the holder level. If the holder was full it would cover for that furnace being off for an hour, that is assuming that the gas use goes on at the normal rate. With the increase in furnace size, however, the time cover has been reduced and provided one good argument for the introduction of the central fuel control described later in this paper.

Owing to the dust content of the gas, considerable concern was felt with reference to the following factors:

(*a*) Accumulation of dust in the holder bottom. This was overcome by fitting special sludging pipes so that the bottom can be periodically washed out.

(*b*) The serious effect of the "sludging" of the sealant and also of its de-watering properties.

These conditions call for specially close attention to water separation in the first settling chambers, and also the maintenance of a steady flow of sealant to the pump chambers.

The tar-sealing medium used is to the following specification:

Specific gravity at 60° F.	1·12–1·125
Water content	Not more than 0·5%
Engler distillation ...	2% at 250° C.; approximately 60% at 350° C.
Fluidity	Liquid at 10° C. for 2 hr.
Viscosity	4–5° Engler at 50° C.; 120–160 sec. at 50° C., Redwood No. 1.

With continual use the sealant liquid becomes emulsified, and a thinning medium is added to bring the viscosity to a reasonable figure. The thinning medium is of the following specification:

Specific gravity at 60° F.	1·100
Water content	Not more than 0·1%
Engler distillation ...	5% at 250° C.; approximately 65% at 350° C.
Viscosity	1·5° to 2·0° Engler at 50° C. 40 to 60 sec. at 50° C., Redwood No. 1.

The usual method of adding the thinning medium is, first, to remove as much of the emulsified sealant as possible, then add a predetermined quantity of thinning oils to the second settling chambers and pump chambers, with good stirring. When the less-viscous material reaches the piston seal, the fluid present in this seal should be agitated thoroughly by means of a smooth, round wood paddle, in order to get a uniform fluid throughout the system as quickly as possible.

GAS MAINS

The usual procedure is to have mains of steel plates of either riveted or welded construction supported on latticed steel stanchions. Explosion doors are fitted along the top of the main and hand-railing is provided in order to make it safe to attend to the explosion doors as well as facilitate the cleaning of the mains. In the semi-cleaned-gas mains, holes have been cut into the top of the main for washing-out, and discharge points have been provided in the bottom of the main at 20 ft. 0 in. centres.

Water seals are fitted at all points considered necessary, particularly at the lowest point of any rise

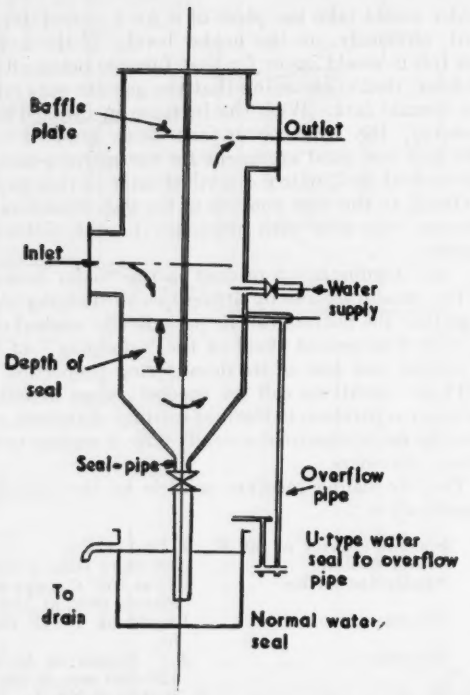


Fig. 2—Diagrammatic arrangement of water-sealed valve providing for difference in level of mains. The baffle plate must be gas-tight at joints to side of walls of box

or fall in the main and at changes of direction. Great care should be taken to ensure that at a change of direction adequate provision is made for clearing the dust deposited.

When deciding upon the velocity of flow it is essential to take into consideration dust quantities and pressure drop along the main. The usual velocities for different conditions are as follows :

Existing Wet-Cleaning System	
Crude-gas systems	15 ft./sec.
Semi-cleaned-gas	20–25 ft./sec.
Clean gas	25–30 ft./sec.
Electrostatic Cleaning Plant	
After dustcatchers	25–30 ft./sec.
Clean gas	25–30 ft./sec.

The two conditions above, it will be noted, do not agree, but it is considered that in the first case, since the gas is burnt in the semi-cleaned state at the stoves, boilers, and soakers, it is better practice to deposit in the mains than block up burners and recuperators. In the case of the electrostatic plant it is better to keep velocities high in the mains before the cleaning plant, in order to attempt to maintain the dust in suspension until it reaches the cleaning plant ; however, it is essential to ensure that in fixing higher velocities the pressure drop along the main is not too great and that adequate facilities are provided to keep the main clean.

INTERNAL CLEANING OF MAINS AND CLEANING SYSTEM

As mentioned previously holes are cut in the gas mains for flushing-out. The usual procedure adopted is to take pressure checks throughout the system at least once a week, in order to locate any serious build-up of dust. A cleaning-gang is, however, permanently engaged in systematically flushing-out from one end to the other. Dust-content tests taken in the common main once a week, and snap checks on each furnace cross main, give a general guide to the dust deposition.

For an example of the type of information taken during pressure checks, in addition to the pressure recorders situated at the gas-holder, boilers, common main, gas washers, and **U**-tube readings the full length of the common main, the following readings are taken at each furnace.

(*a*) Pressure at washing-tower inlet.
(*b*) Pressure at washing-tower outlet.
(*c*) Pressure at fan inlet.
(*d*) Pressure at fan outlet.
(*e*) Pressure at cross-main inlet.
(*f*) Pressure at entrance to 8-ft. main.
(*g*) Pressure at 8-ft. common main.
(*h*) Fan load (amperes).
(*i*) Gas flow (cubic feet per hour).

From the information compiled an accurate picture of the condition can be ascertained, and action can be taken to clean the mains and system before any serious delays are caused.

PURGING OF MAINS AND CLEANING SYSTEM

It is vital to provide steam connections and purging-cocks to all gas mains and sections of the cleaning system, in order that the system between the isolating valves may be gassed systematically. The steam connection is placed close up to the valve on the downstream side, and the purging cocks on the upstream side. When gassing any section a bank of steam must always be provided before the gas stream, and the section must be full of gas before closing the purging-cocks.

ISOLATING VALVES

(*a*) Water-Seals

Water-seals are used to isolate any section of the distribution system, and usually incorporate a change in direction or in level of the main. Typical arrangements are given in Figs. 2 and 3. It must be made clear that this type of valve is not used near the furnaces, but as mentioned above. The water-seal is very simple in design, and the cost is negligible in comparison with the goggle and dumpling valves

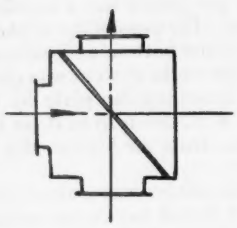


Fig. 3—Plan of water-sealed valve providing for change of direction. Seals arrangement as in Fig. 2

and, since during operation a continual flow of water is maintained, safety precautions are adequate. The main advantages are, firstly, mechanical maintenance is nil and, secondly, with a semi-cleaned-gas system a further water seal and cleaning point is provided.

The valve consists of a steel box with an internal baffle plate separating the inlet and outlet. From the bottom of the box is taken a seal-pipe to a normal water seal, and into the side of the box a water supply and overflow pipe which is sealed by a U-tube water seal. The distance from the overflow pipe to the bottom edge of the baffle plate is the depth of the seal, which is usually fixed to overcome a pressure at least double the normal working pressure. When the valve needs to be sealed, the valve on the seal pipe is shut and the box filled up, the water overflowing through the overflow pipe. It is imperative that a constant watch be kept on this overflow pipe to ensure that water is continually flowing. For normal running the seal-pipe valve is opened and the box used as a water seal.

***(b)* Dumpling Valve**

With the existing system the dumpling type of valve is introduced at the cross-main inlet to the common main and isolates the furnace, an arrangement being shown in Fig. 4. The valve is of the simple-lift type, with a valve-tube mounted on the central spindle. The seat is formed by a circular trough into which the valve-tube lip enters. During service this trough fills with dust, and on the rare occasions that the valve is closed the valve-tube lip penetrates the dust and acts as a seal. This valve, however, does not make a perfect gas-tight joint, so that when the furnace needs to be isolated the usual practice is to introduce a blank immediately under the valve.

The valve casing is constructed of mild-steel plates and is of sufficient height that the valve in its top position is out of the gas flow. The valve is operated through chains and chain pulleys and is balanced for easy operation.

***(c)* Goggle Valves**

For the conversion to electrostatic filter plant it is intended to isolate each furnace by a goggle-type valve immediately after the dustcatcher. In fact, on a recent rebuild a thermal-expansion-type goggle valve was introduced, and this valve is described below and a sketch is shown in Fig. 5.

Goggle valves provide a reliable gas-tight joint, and in principle introduce a blank into the main. The valve consists of a goggle plate which is so operated that when the valve is open the hole in the plate registers with the tube diameter, but when closed the blank portion shuts off the main. A goggle plate release device is fitted to allow the plate to slide freely and is attained in the thermal-expansion type by means of the thermal expansion of metal tubes so as to open the flanges in which the goggle plate moves. To take up this flange movement an expansion joint is provided in the main.

Three expansion tubes are equally spaced around the valve and fitted with pull rods. These rods are securely fastened to the anchor flange at one end and

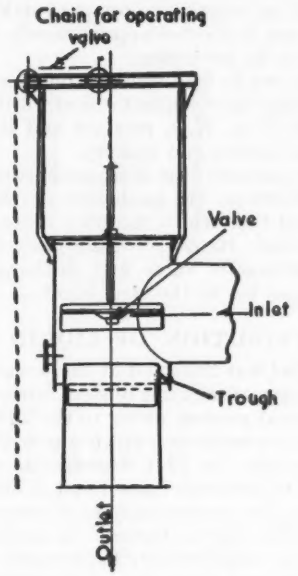


Fig. 4—Diagrammatic arrangement of dumpling valve

to the expansion tube at the other by means of a cross-head and adjusting nuts. The initial adjustment of the flange against the goggle plate is done by means of these adjusting nuts. When steam is admitted to the tubes a force is produced by the expansion of the tubes, which, since one end of the tube is fixed by pull rods to the rod anchor flange, produces movement to the other flange to which the expansion tubes are anchored.

The plate is swung by worm-gear unit and an operating chain attached to the plate. After the plate is swung the steam is turned off and the tubes cooled by admitting water. This produces contraction of the tubes, with a subsequent movement of the flanges together, clamping the plate.

DISTRIBUTION OF COKE-OVEN GAS

Coke-oven gas does not present the problems met with in semi-cleaned blast-furnace gas, but the general details of mains follow in principle those of clean blast-furnace gas, particularly in regard to steam-purging which is still essential for safety reasons. The mains are of welded construction and the velocity of the gas is usually in the region of 25 ft./sec.

The gas is boosted to the coke-oven-gas-holder, which is a single-lift spiral-guided type of 100,000-cu. ft. capacity and throws a pressure of 12 in. W.G. when full. A diagrammatic arrangement of the distribution is shown in Fig. 1. The main distribution centre is at the outlet of the gas-holder, and each offtake is fitted with an automatic valve. The sinter-plant leg delivers to a gas-holder, approximately 1000 yd. distant, of the single-lift column-guided type of 20,000 cu. ft. capacity and throwing 7 in. W.G. pressure when full. The gas to Nitrogen Fertilisers,

Ltd., which is a separate company taking surplus coke-oven gas, is boosted approximately 2 miles into a 340,000-cu. ft. gas-holder.

The main works feed splits at the steelplant end; one leg feeding the steelplant is controlled by Askania regulator to 7 in. W.G. pressure and the other leg supplies the boilers and soakers.

The rest-gas main from Nitrogen Fertilisers delivers into a 100,000-cu. ft. gas-holder of the single-lift spiral-guided type which throws a pressure of 12 in. W.G. when full; the outlet of this gas-holder is fitted with an automatic valve and discharges into the coke-oven-gas leg to the steel plant.

DISTRIBUTION OF LIQUID FUEL

Liquid fuel was first used at the works in 1939, in the form of crude-tar firing of steel furnaces, without, however, great success, owing to the high proportion of solid matter in the tar, which was used direct from the coke ovens. In 1941 experiments were carried out at the boilers with creosote-pitch firing, in order to ascertain the correct mixture of creosote to pitch and a suitable type of burner. It was found that a 50/50 mixture was necessary to overcome the settling-out of solids, and the design of burner is still in use to-day with only slight modifications, and is described later in this paper. Two steel furnaces and two mixers were then converted to creosote-pitch firing. During 1946, when the fuel situation was becoming desperate, a further system was installed for burning heavy Pool fuel oil at three steel furnaces.

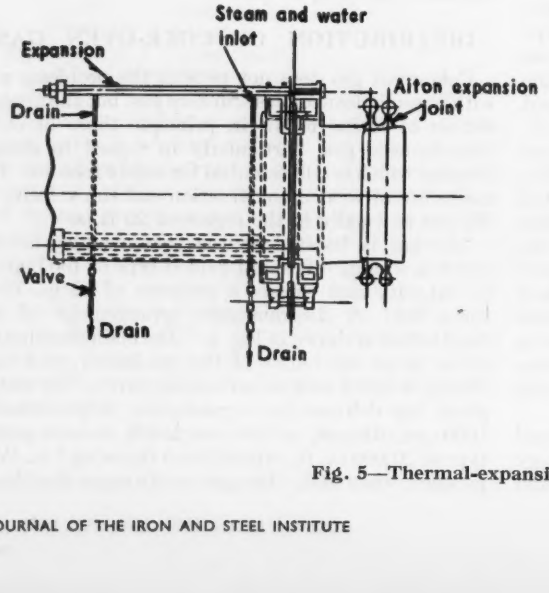
These installations were built with materials that were readily available and have in their present state certain drawbacks. Therefore, it is intended to describe the new plant, which is being installed as quickly as delivery of materials will allow. Details are included which, it is hoped, will be useful to the engineers responsible for the design and maintenance of such plants. A diagrammatic arrangement of the layout is shown in Fig. 6.

The approximate characteristics of the two liquid fuels at present in use are as follows:

	Heavy Pool Fuel Oil	Creosote-Pitch
Composition (ultimate):		
Carbon, %	85·5	89·86
Hydrogen, %	11·2	6·05
Sulphur, %	2·1	0·39
Specific gravity at 15·5° C....	0·95	1·16
Kinematic viscosity, c.g.s. units (Stokes):		
At 80° F....	7·0	13
At 100° F.	3·0	3·9
At 120° F.	1·28	1·5
At 140° F.	0·88	0·69
At 160° F.	0·5	0·37
At 180° F.	0·35	0·22
At 200° F.	0·24	0·13
Calorific value, gross B.Th.U./lb.	18,500	16,250
Flash point (Pensky-Martens), ° F.	230	150

UNLOADING BAY

The fuel oil is delivered in tank wagons, and therefore the first essential is an adequate unloading bay with unloading pumps and steam heating for the tank wagons. The installation shown consists of ten 4-in. connections—two per wagon, spaced at 21 ft. 0 in. centres—into a 6-in. bore header-pipe running in a concrete duct between two railroads and delivering into the unloading pump suction. A 2-in. bore steam pipe is to be clipped to the header pipe, and ten 1-in. dual connections spaced as above are provided for steaming tank wagons. Dust caps and chains are to be provided for the 4-in. connections.



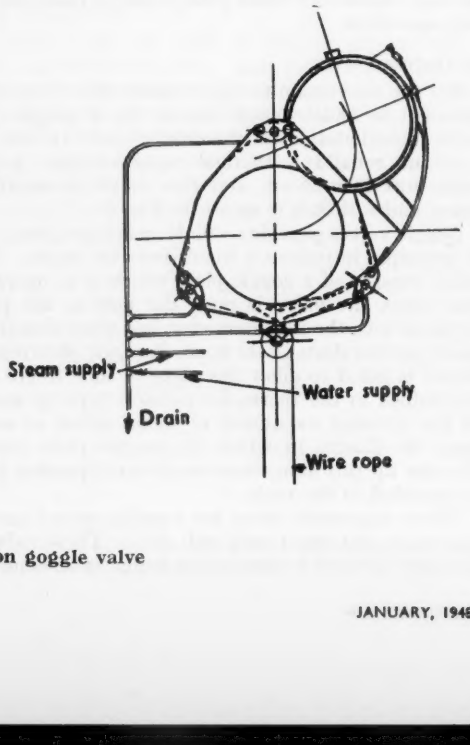


Fig. 5—Thermal-expansion goggle valve

There are two unloading pumps of the horizontal duplex type, 8 × 8 × 8 in., each capable of delivering 9000 gal./hr. against 50 lb./sq. in. total head, with a steam pressure of 150 lb./sq. in. They are fitted with steam and exhaust stop valves, air vessel, and a relief valve with delivery into the pump suction. They discharge into either of two storage tanks. A coarse filter will be installed on the suction leg of these pumps.

STORAGE TANKS

The essential storage requirements are as follows:

(1) An adequate stock of fuel should be carried to guard against possible interruption of supplies. The two tanks being installed are each of 370 tons capacity; this is sufficient to supply the steel plant for a fortnight.

(2) The fuel should be stored at such a temperature that it may readily be fed to the pumping system at the desired maximum rate. On the other hand the fuel should not be at too high a temperature or sedimentation and partial evaporation will occur, with consequent difficulties.

The heater consists of steam coils and should be constructed with best-quality steam tubing. All joints immersed in the fuel should be welded. Coils should be arranged so that they drain freely to a steam trap. The usual source of steam supply is the waste-heat boilers, and as the pressure is too high, a reducing valve is fitted. Oil-temperature control will be maintained by manual operation of the steam supply to the coils, in accordance with limits scribed on the chart of a temperature recorder operated from a mercury thermometer.

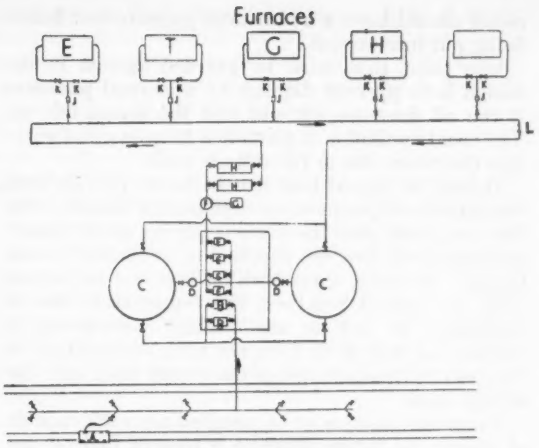
In the existing installation the oil is brought to its atomization temperature in the storage tank. The temperature and corresponding viscosities are as follows:

Creosote-pitch 160° F. = Kinematic viscosity 0·37 Stokes = 124 sec., Redwood No. 1.

Heavy Pool fuel oil 170° F. = Kinematic viscosity 0·425 Stokes = 160 sec., Redwood No. 1.

However, in the new installation the heavy Pool fuel oil in the storage tanks will be heated only sufficiently to give a reasonably free pump circulation and reduce cavitation in the pump. It is intended to heat to a temperature of approximately 100° F., corresponding to a kinematic viscosity of 3 Stokes. The final heating will then be done by flow heaters in the delivery leg of the supply pumps.

The tanks are of welded construction, and are fitted with manholes, sludge cock, and air vents. Level recorders are also incorporated. One point that must be watched when using both coal-tar and petroleum fuels on different installations is that the fuels are not mixed; otherwise changes in physical state occur and render the mixture unfit for use as a fuel. When changing from one fuel to the other, as so often happens, it is essential that the system be first cleaned-out with creosote. The method adopted is to run the storage tank as low as possible, fill it with approximately 24 tons of creosote, and then pump this through the system, burning the fuel at the furnace to ensure that all the mains are clean.



A Tank wagons (12–14 tons capacity)
B Unloading pumps (two)
C Storage tanks (two)
D Filter
E Supply pumps (four)
F Large-capacity air vessel
G Air compressor for charging air vessel
H Flow heaters (two)
J Orifice plate and carrier ring
K Reversing valves
L Relief valve

Fig. 6—Diagrammatic arrangement of fuel-oil installation

FUEL TRANSFER

When possible, in order to stabilize conditions within the mains system, it is desirable to have in circulation at least three times the quantity of oil required at the furnaces. Further, when using direct-acting pumps delivering through long mains, it is usual to fit an air vessel as near to the pump as possible, to minimize variations in flow and to damp-out pressure pulsations. Air vessels have been used on the existing installation, but unless they are fitted with a small air compressor in order to replenish the air volume as it is absorbed by the oil they are of very little use. The control of the air compressor is usually by float arrangement in the air vessel to maintain the air volume at 30–40 times the oil volume.

There are four supply pumps to be installed, each capable of delivering 2000 gal./hr. against 50 lb./sq. in. total head with steam pressure of 150 lb./sq. in. They are of the horizontal duplex type, $4\frac{1}{2} \times 3\frac{3}{4} \times 4$ in., each complete with stop valves, air vessel, and relief valve discharging into the pump suction.

FLOW HEATERS

As mentioned previously, flow heaters are not used, since the oil is heated to its atomization temperature in the storage tanks, but the usual method and the one to be used on the new installation is to provide flow heaters in the delivery leg of the supply pumps. These heaters are to be in duplicate, each of sufficient capacity to carry the normal load to permit cleaning and inspection without interrupting service.

Heaters should be examined regularly to detect leaks between the oil and steam passages, and after

repair should have a hydrostatic pressure test before being put into service.

One point that must be guarded against in the design is to prevent damage by abnormal pressures if the oil lines are cut out and the steam left on. The usual method is to provide a by-pass relief valve into the return line or the storage tank.

It may be argued that it is better to provide final temperature adjustment on each burner branch ; this however, multiplies the items requiring maintenance, particularly if they are supplied in duplicate to each burner. Further, since both oil legs and the steam main are lagged together, the temperature loss is negligible ; in fact on the existing installation, it amounts to only 3° C. from the tank temperature to the delivery temperature of the return main into the storage tank.

The steam heaters to be installed are each capable of raising the temperature of 8 tons of fuel oil per hour from 80° F. to 220° F., when supplied with steam at between 80 to 200 lb./sq. in. The shell is of welded mild steel, the cover of cast iron, and the tubes are U-shaped of seamless drawn steel expanded into a mild-steel tube-plate with mild-steel shaped flow plates. The mountings comprise a thermostatic valve and a steaming-out valve, a relief valve, and a dial thermometer with pocket. The shell is lagged with block magnesia and cleated with steel sheet.

The oil temperature will be controlled to approximately 195° F., corresponding to a kinematic viscosity of 0·25 Stokes. However, since the heaters will be in the delivery leg of the pump the temperature may be increased to any limit as long as vaporization does not occur, so that the final temperature arrived at after experience is gained cannot be stated at the present time.

FILTERS

All fuel installations should include duplicate filters interconnected so that one may be cleaned at any time. They should be placed preferably on the suction side of the pump. Pressure gauges should be fitted on both sides of the filter to indicate when a change-over is necessary. Individual filters have been tried in the burner legs, but they require too much attention and, as in the case of the flow heaters, it is considered better practice to instal the duplicate filters at the pumping installation.

MAINS LAYOUT

The ring-main system is the type usually adopted, and the essential requirements are as follows.

(1) The size of the main must be fixed so that the pressure drop due to friction is not excessive.

(2) The supply leg must rise to a point past the last offtake and the return leg fall to the storage tank, so that the main automatically drains.

(3) A relief valve is incorporated in the main in order to maintain the pressure, and is usually located at the high point in the main.

(4) In order to simplify erection and lagging and to minimize heat loss, it is better practice to run the supply and return oil mains and the steam main together. However, if this is not possible it is essential to provide steam tracer lines to all oil mains.

(5) For efficient control of the furnace and to facilitate quick reversal it is imperative to have a central control panel, with the mains distributed from this point to each end of the furnace. A description of this panel is given later in the paper.

(6) The distribution valves on the panel should be duplicated so that valve change does not delay the furnace.

(7) A by-pass should be provided for the oil-measurement orifice plate so that no delay is caused by maintenance required or in the event of the fuel being changed for any reason.

Details of the furnace distribution are given later in this paper.

DISTRIBUTION CONTROL

GENERAL CONSIDERATIONS

Owing to improvements in blast-furnace design and to developments in ore beneficiation, there has been a very considerable reduction in the amount of coke required in the blast-furnace per ton of pig iron produced, with a subsequent reduction of the amount of blast-furnace and coke-oven gases available. Further, due to the shortage of fuel supplies, particularly in the form of solid fuels, more reductions have taken place in the amount of producer gas and coke-oven gas available for the normal heat requirements. Therefore, in order to satisfy the heat requirements of the works, it was found necessary to import auxiliary fuels and electrical power. The problem of heat distribution has thus become much more complicated than was the case a number of years ago.

However, before the shortage of fuel assumed anything like its present state, when the operation of the works depended largely on the supply of blast-furnace and coke-oven gases, it was decided that a very good case could be made out for some method of central fuel control, where all the information in connection with the make and distribution of gaseous fuels could be centralized in one room and the problems of re-allocation could be dealt with without walking round

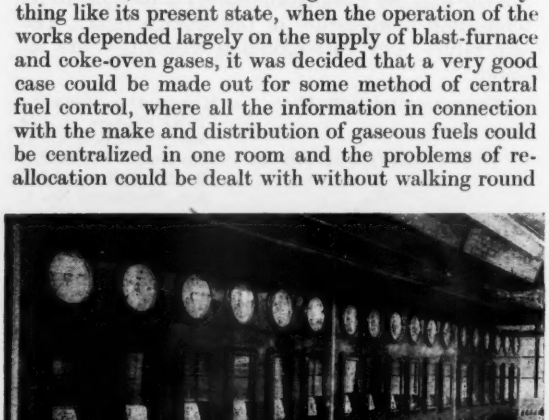


Fig. 7—Fuel-control-room panel

the works to find out the true position. A fuel-control room was accordingly built, and was completed in 1945. A description of this control room, together with the method of control, is given in the following sections of this paper.

FUEL-CONTROL ROOM

The fuel-control room is equipped with a main instrument panel, an internal telephone system, and gas-holder-position indicators. A photograph of the control room is shown in Fig. 7.

(a) *Main Instrument Panel*—The main instrument panel gives the following information :

(i) Blast-furnace-gas make at No. 1 blast-furnace.
Blast-furnace-gas make at No. 2 blast-furnace.
Blast-furnace-gas make at No. 3 blast-furnace.
Blast-furnace-gas make at No. 4 blast-furnace.
Blast-furnace-gas make at No. 5 blast-furnace.

(ii) Blast-furnace-gas use at steelplant.
Blast-furnace-gas use at coke ovens.
Blast-furnace-gas use at soaking pits.
Blast-furnace-gas use at boilers.

(iii) Coke-oven gas to gas-holder.

(iv) Coke-oven-gas use at coke ovens.
Coke-oven-gas use at steelplant.
Coke-oven-gas use at soaking pits.
Coke-oven-gas use at sinter plant.
Coke-oven-gas use at Nitrogen Fertilisers.
Coke-oven-gas use at boilers.

(v) Return (or rest) gas from Nitrogen Fertilisers to steelplant.

There are 18 sets of instruments on the panel, consisting of an indicator, recorder, and integrator operated by an electrical transmitter from Kent ring-balance-type meters working from a differential across an orifice plate in the gas mains.

(b) *The Telephone System*—The telephone system consists of a self-contained unit, with the switchboard on the control-room-operator's desk. The phones are located at the various points in the works where it is necessary to receive or give information.

(c) *Gas-Holder-Position Indicators* — Gas-holder-position indicators were found necessary in order to have a knowledge of the amount of gas available in the coke-oven- and blast-furnace-gas-holders. A system of lights was therefore installed on the operator's desk so that he could see the position at a glance.

For the blast-furnace-gas-holder the lights indicating the position are graduated at 250,000-cu. ft. intervals, and cover a range from 500,000 to 2,000,000 cu. ft., and for the coke-oven-gas-holder the bottom light indicates that an automatic valve in the holder delivery is shutting and the top light indicates bleeding, while intermediate lights indicate 30–60,000 and 60–90,000 cu. ft. While this system meets essential points, it is admitted that the controller at any particular time is not aware whether the holders are rising or falling until a light changes. In the case of the blast-furnace gas-holder this drawback is overcome during the times of gas shortage by the holder attendant keeping in close contact with the controller. In addition, the gas-holder-pressure recorder is mounted so that the controller can keep a fairly accurate check. The coke-oven gas presents a different problem but is well covered, since Nitrogen Fertilisers take surplus gas and regulate their gas booster in accordance with the holder position.

Owing to the existing layout of the gas mains it was found impossible to meter the blast-furnace stoves, the old boiler plant, and the power-house, all taking blast-furnace gas. However, these difficulties are overcome by telephonic communication of any alterations required, but it is intended, in the alterations to the gas distribution necessitated by the installation of an electrostatic cleaning plant, to have all uses metered. Further developments envisaged are the introduction of control to the make and distribution of steam, since this is also a problem which is vital in the efficient running of iron and steel works.

METHOD OF CONTROL

It was decided that the control room should be manned continuously for a seven-day week, and as the chance of obtaining the services of experienced fuel engineers was remote, it was decided to train personnel from the works who had a good idea of the general layout. These men were given a practical tuition in the duties involved, and also a period of training at the works.

The function of the control room is to ensure that full use is being made of the blast-furnace and coke-oven gases available, and that auxiliary fuels are being used in the required amounts to ensure that the works are being supplied with the necessary fuels to give the desired operating conditions.

The auxiliary fuels at present in use are coke breeze, producer gas, and creosote-pitch, as well as electrical power.

(a) *Coke Breeze*—The coke breeze is obtained from the screening of coke, and is used on two 30,000-lb./hr. boilers. It may be used at the rate of 1–4 tons/hr., according to the blast-furnace-gas position. The equivalent blast-furnace-gas value is 200,000 cu. ft./ton.

(b) *Producer Gas*—The producer gas is obtained from two gas producers and may be used on one steel mixer and two steel furnaces. The producers have a maximum rate of 6 tons/hr. of deep-mined producer coal, or 4 tons/hr. of outcrop coal. Producer gas replaces blast-furnace and coke-oven gases.

(c) *Creosote-Pitch*—Creosote-pitch may be used on the steelplant mixers and replaces coke-oven gas to the extent of 20,000–30,000 cu. ft./hr.

(d) *Electrical Power*—Electrical power is imported from the grid system. The quantity varies according to the gas available for works' generation and normally fluctuates to the extent of about 1000 kW. In the case of emergency, electrical equipment is available for the grid system to take the full works' load.

The amount of auxiliary fuel used can thus be varied to meet plant requirements, giving complete flexibility and ensuring that the plant is maintained at full output, unless conditions are so serious that this is impossible.

For really efficient control it is essential that departments should inform the controller of any modification, so that neutralizing action can be taken. For example, since the rolling programme directly affects the steam demand, the mill foreman contacts the control room at the beginning of his shift and details, as far as possible, the shift programme. From experience, the controller and the boiler attendant have a good idea of the steam requirements, and the boilers and the works' generation are set accordingly.

When the control room was first put into operation, the tendency of all departments was rather to mistrust it, but it was soon realized that the link between departments, which previously, to a certain extent, had been missing, could be provided by this control room to the benefit of all. One essential factor is that the controller should be very tactful in dealing with departments, or the whole system becomes unmanageable. If he has to modify the fuel allocation, the greatest care must be taken to inform the department affected of the reason why the alteration has to be made.

FUEL UTILIZATION

The essential requirements on an integrated iron and steel works are firstly to utilize the by-product gases to the best advantage, taking into consideration the temperatures to be attained and metallurgical factors, secondly, to consider the necessity of making provision for the use of additional fuels to meet the normal heat requirements, and thirdly the use of auxiliary fuels to replace the primary fuels when in short supply, in order to maintain full output.

The utilization of by-product gases can be discussed separately for the services for which they are, or can be, employed, but consideration must also be given to the priority of these different services for the gas supply available. The hot-blast stoves, for instance, will certainly be fired with blast-furnace gas, since the gas is in every way suitable for this purpose and is the fuel most readily available. In the case of our plant, where the coke ovens are adjacent to the blast-furnaces, blast-furnace gas should certainly be used for underfiring the ovens ; in fact there are advantages in its use, for the following reasons :

(1) Smooth and even heating of an oven wall can be much more readily obtained.

(2) The carbonizing temperatures are such that with the regenerative facilities provided they can be maintained by combustion of lean gas.

(3) The use of blast-furnace gas releases coke-oven gas for metallurgical processes, and blast-furnace gas can be mixed with this in suitable proportions for melting furnaces, soakers, etc., or the coke-oven gas may be used with liquid fuel on melting furnaces or used direct, as at the sinter plant.

There remain the questions of blast-furnace blowing and power production. If gas engines are used, they must take precedence over almost any other service, since if the gas is not available they will come to a standstill. It is a real advantage of the steam system that boilers can be fired with alternative fuels as efficiently as with blast-furnace gas, and that gas and, for example, coke breeze, can be used together in any proportion that may be desirable. The boiler plant, therefore, has a real value as a cushion to absorb temporary variations in the balance between gas supply and consumption.

In the following paragraphs will be found descriptions of the existing plant and the modifications to be included in the rebuilding programme.

BLAST-FURNACE STOVES

The existing blast-furnace stoves are of the two-pass type, 21 ft. 1 in. in dia. and 85 ft. 0 in. in overall height. The combustion chamber is elliptical and has a cross-sectional area of 29¼ sq. ft. The checkerwork, the total height of which is 70 ft. 8 in., is constructed of 12 × 5 × 2½-in. bricks, the size of the checker openings being 5 × 5 in. square, and the total sectional area of the checker passages is 94 sq. ft. The weight of the checkerwork is 390 tons, and the total heating surface is 64,000 sq. ft.

The burner is a parallel tube 2 ft. 0 in. in dia. leading into the bottom of the flame flue. An isolating valve of the spectacle type is fitted, and the gas flow is controlled by a butterfly valve in the vertical connection between the 8-ft. gas main and the burner. This connection is so short that an orifice plate for gas-flow measurement could not be installed. The air combustion is supplied by two openings in the stove side, one on either side of the gas inlet, the amount of air being governed by a pivoted cover-plate. When using semi-cleaned gas the efficiency of these stoves is in the region of 75%.

For the conversion to clean-gas firing the stoves will be relined, decreasing the checker openings, and new pressure burners with automatic air/gas ratio control will be installed. Automatic blast-temperature control will also be fitted.

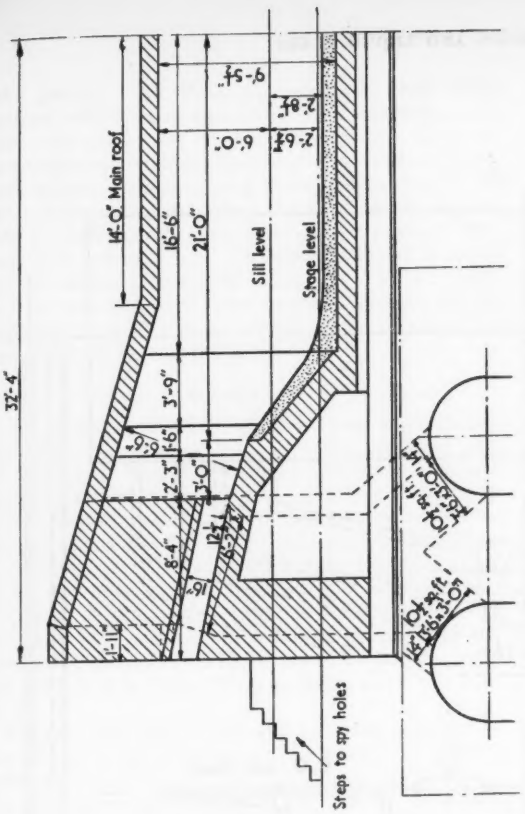
COKE OVENS

The existing coke-oven plant consists of two batteries of the Becker type, comprising 70 ovens in all. The first battery, of 47 ovens, was installed and fired in 1932, and the second battery, of 23 ovens, was fired in August, 1938. Both batteries have continued in service without any stoppage.

As mentioned previously, the most economical gas to use is blast-furnace gas, and the only problems from the engineer's point of view are to maintain a constant gas pressure, and to ensure that the dust content under any circumstances does not exceed 0·02 g./cu. m. The most important point is to maintain regular running of the plant, and with this in view the coke ovens are the last department to be affected by any shortage in gas supply. Granted, if the supply position is serious, the change-over to coke-oven gas underfiring is not delayed under any circumstances.

BOILERS

The existing boiler plant consists of three 30,000-lb. Babcock and Wilcox water-tube boilers, two of which are fitted with mechanical chain-grate stokers burning



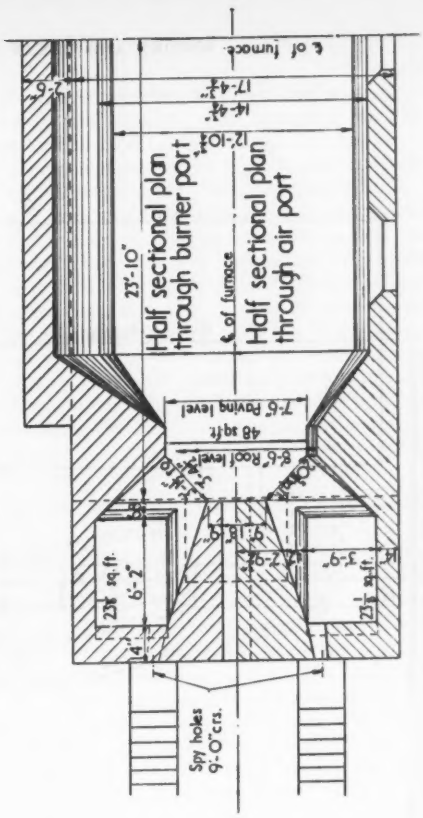
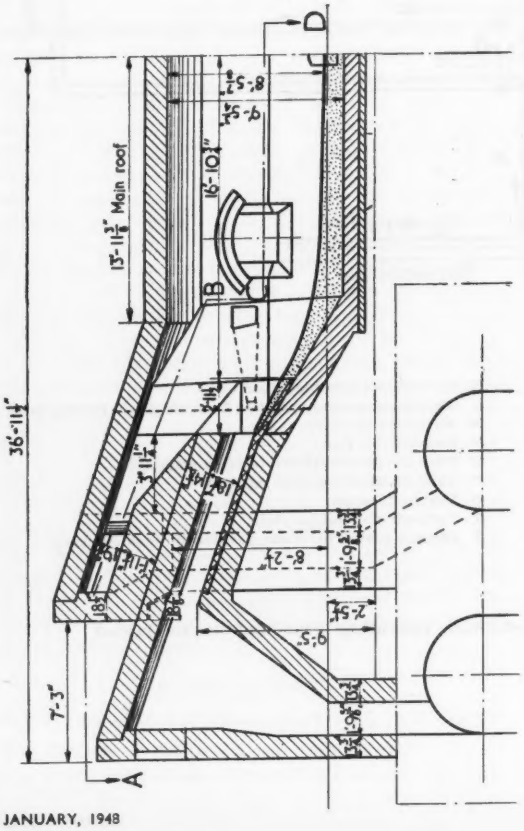


Fig. 9—Diagrammatic arrangement of 120-ton furnace *A*.
Liquid fuel and coke-oven-gas fired



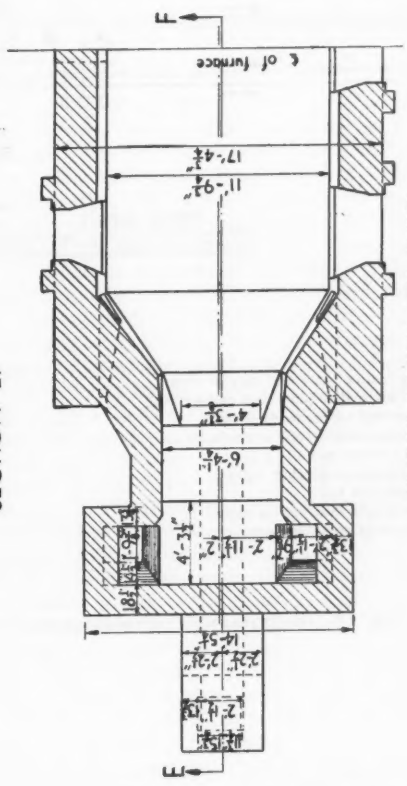
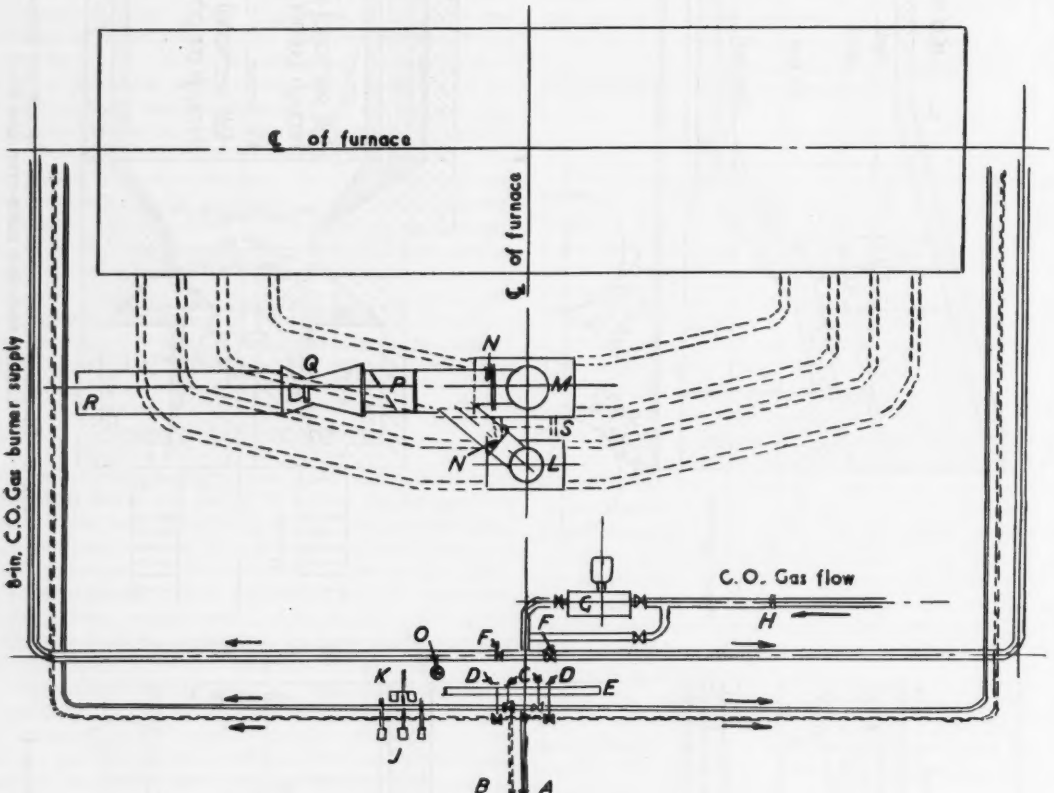


Fig. 8—Diagrammatic arrangement of 65-ton fixed furnace *A*.
Gas fired



A Oil-flow orifice plate and carrier ring
B Steam-flow orifice plate and carrier ring
C Oil-reversing valves in duplicate
D Steam-reversing valves in duplicate
E Central control panel
F Coke-oven-gas reversing valves
G Coke-oven-gas fan
H Coke-oven-gas-flow orifice plate
J Furnace-door operating gear
K Air-reversal operating gear
L Air-reversing valve modified from old gas-reversing valve
M Air-reversing valve
N Butterfly air flow
O Total-air-flow-regulator valve on stage level
P Total-air-flow butterfly regulator
Q Venturi-tube fan
R Total-air-flow orifice plate
S Damper control for waste-gas flow through checkers

Fig. 10—Diagrammatic arrangement of steel-furnace operating gear and mains layout

coke breeze. All three are equipped with Gako burners, which are designed to burn blast-furnace and coke-oven gas in any desired proportions. There are also thirteen 10,000-lb. Babcock and Wilcox water-tube boilers with gas firing in any proportions. All these boilers are generating steam at 150 lb./sq. in., 480° F., and the efficiencies are in the region of 75% and 65%, respectively. The installation of a steam system working on a 450 lb./sq. in., 750° F., cycle, and the meeting of the low-pressure demand by the existing three 30,000-lb. boilers and the steelplant waste-heat boilers is being considered.

SOAKERS

The existing plant consists of four regenerative pits, four recuperator pits, and six pressure pits, all fired with blast-furnace and coke-oven gas as a boost, in different proportions according to the type of pit. As automatic control is not fitted to any of the pits, it is considered desirable that a brief description should be given of the new soaking pits being installed. These are one-way-fired soaking pits of Stein and Atkinson design, and consist of three batteries of four pits each. Each pit will hold eighteen 3½-ton ingots approximately 21½ in. square by 6 ft. long, and is 25 ft. 0 in. long, 7 ft. 0 in. wide, and 10 ft. 3 in. deep.

The new pits will be arranged for firing with mixed blast-furnace and coke-oven gas having a calorific value of 135 B.Th.U./cu. ft. The mixing is controlled by an Askania regulator arranged for panel mounting. The total air for combustion will be preheated in Stein refractory type recuperators and will be delivered to the burners by a hot-air fan of special design suitable for a maximum air preheat of 1100° F.

Complete automatic control is to be provided. Temperature control of each pit will be by Kent Multelec recording controllers, giving floating proportional control. The controller operates from a thermocouple in the front of the pit and actuates a motorized gas valve in the main of each burner. Fan-air temperatures will be controlled in order to provide against the fan overheating. A cold-air valve is to be provided, operated by a Kent proportioning controller actuated from a thermocouple in the hot-air connection. Combustion control will be provided by an Askania regulator operating from differential pressure across orifice plates in the gas and air mains and adjusting the air flow to maintain a predetermined air/gas ratio.

The pit pressure will be controlled by an Askania regulator operated from a connection from the end wall of the pit. The regulator will adjust the damper in the flue between the pit and recuperator.

The Kent meters and Askania regulators will be arranged for flush panel mounting and one plate panel will be provided for each pit. The four panels for each battery of pits are to be mounted in a glass-fronted control house.

STEELPLANT

COKE-OVEN-GAS/LIQUID-FUEL FIRED FURNACES

The furnace to be described was originally 65-ton fixed, fired with blast-furnace and coke-oven gas, and for the conversion to liquid-fuel and coke-oven-gas firing the capacity was increased to 120 tons. Details of the modification to the furnace top will be seen from a comparison of Figs. 8 and 9, which show the original and existing lines, respectively.

It will be noted that both the gas and air checkers were utilized for preheating air and were not modified as to location and size, but the openings were increased from 9 × 3 to 9 × 7¾ in. In order to provide for the additional capacity, the port ends were moved out and the main uptakes, in addition to being made as large as possible, were located as central as possible to the slag canals in order to provide greater slag capacity and to ensure that the natural flow of waste gases to both checker chambers would be equal. This flow can, however, be regulated by a damper between the two reversing boxes.

MAINS LAYOUT AND CONTROL GEAR

A diagrammatic arrangement of the layout of the mains and control gear is given in Fig. 10. The coke-oven gas is utilized cold, and is boosted by a fan to a pressure of 25 in. W.G. The suction leg of this fan is taken from the common main, and the gas flow is measured by the differential over an orifice plate operating a ring-balance meter with indicator and recorder. The fan may be by-passed in case of repair, and the coke-oven gas is then utilized at the mains pressure. The delivery from the fan is split to each end of the furnace at a **T**-piece, and each leg is provided with a reversing valve. The **T**-piece is located under stage level immediately in front of the central control panel so that the valves are at approximately 3 ft. centres. The vertical leg at each furnace end is provided with a ball joint to facilitate burner setting.

The oil and steam offtakes from the ring mains are provided with furnace-isolating valves near the ring main, and the flow is measured by the differential across orifice plates operating meters with indicators and recorders. At present new Kent meters for both steam and oil are being installed, and it is hoped that these will solve one of the main problems in the use of liquid fuel, because, although we have flow indicators and recorders on the furnace being described and one other, the remainder are fitted only with integrators at each burner with end control, which is completely unsatisfactory for control efficiency.

In the case of the old gas-fired furnace the air was supplied by a Venturi-tube fan driven by a 4·7 h.p. motor and capable of delivering 12,000 cu. ft./min. against 1 in. W.G. pressure at 975 r.p.m. The air was measured by an orifice plate at the end of the fan suction tube, and the flow was controlled by a butterfly valve in the fan delivery leg, which is operated from stage level. The modifications carried out were to disconnect the gas mains to the gas reversing valve and provide an air connection to this valve, by a breeches pipe in the fan delivery leg. In order to maintain equal flow through the checkers, butterfly valves were installed in each leg to the reversing valves. These reversing valves are hydraulically controlled and the operating levers, at stage level, are connected for ease of operation.

The control panels, at present being installed as

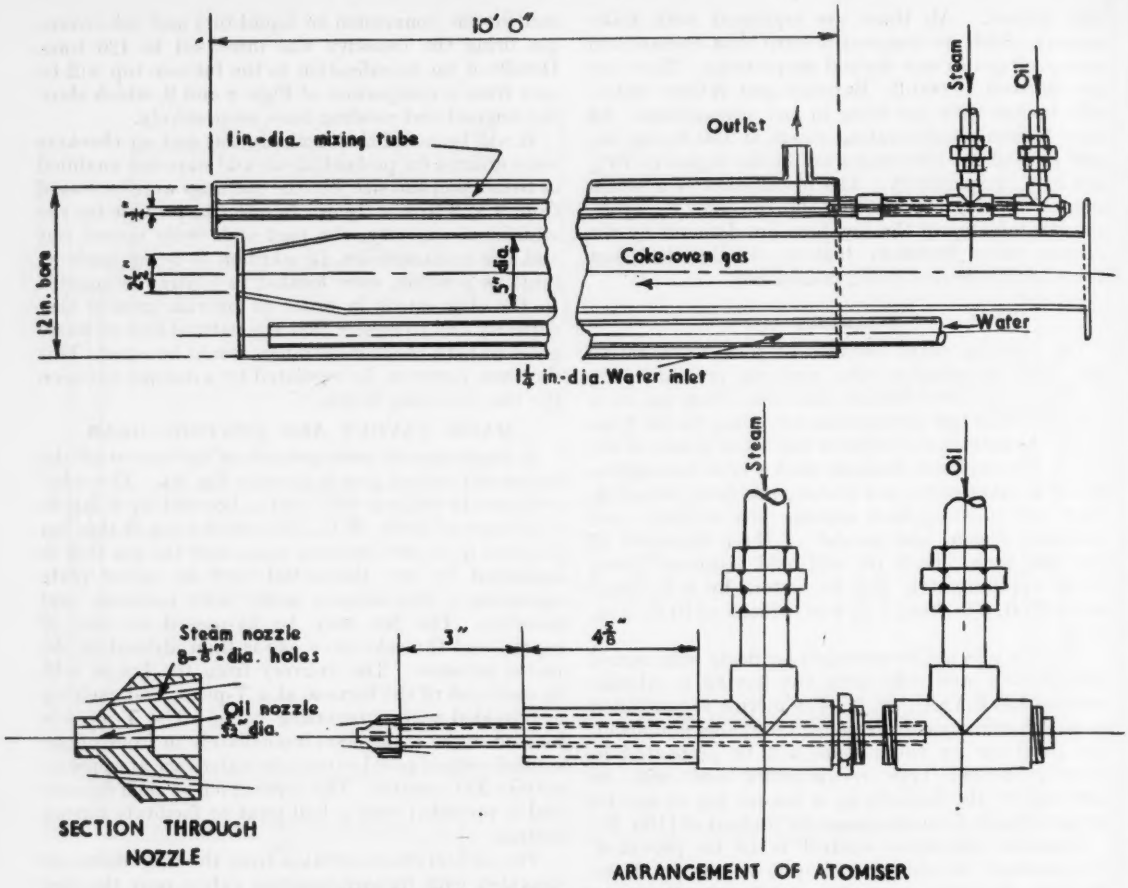


Fig. 11—Coke-oven-gas and liquid-fuel burner

materials become available, include the following instruments :

Gas-flow indicator and recorder.
Air-flow indicator and recorder.
Oil-flow indicator and recorder.
Steam-flow indicator and recorder.
Oil-pressure gauge.
Steam-pressure gauge.
Gas-pressure recorder.
Roof-pressure recorder.
Waste-heat-boiler-suction recorder.
Four flue-suction recorders.
Flue-gas-temperature recorders at waste-heat boiler.

Oil- and steam-control valves are also incorporated on the panel.

BURNERS

The design of burner which has been found to give the best results up to the present time is shown in Fig. 11. These burners are manufactured in our own workshops and are of welded construction. As mentioned previously, the coke-oven gas is boosted to a pressure of 25 in. W.G. at the fan outlet, and the gas tube is nipped down at the burner tip to 2½ in. dia., giving a gas velocity of 250–300 ft./sec.

The atomizer nozzle is of brass and is made a push-fit into the 1-in. tube, and the atomizer is so constructed that this nozzle may be withdrawn at any reversal for examination. Both the oil and steam are used at approximately 30 lb./sq. in. pressure, and the 1-in. tube is sleeved-down at the burner tip to a diameter of ¾ in. One feature incorporated was to cut back the burner end to obviate carbonization of the gas-port nozzle owing to dripping of the oil.

The main steam reversing valves are drilled with a ⅛-in. hole in order to maintain a continual flow of steam through the mixing tube for cleaning purposes.

The burner is water-cooled and requires approximately 1100 gal./hr. Owing to the water-cooling, condensation may take place in the 1-in. mixing tube, and to obviate this it is intended to insulate with slag wool.

The burner described is by no means final, and experiments will be continued in an endeavour to find the most efficient type.

With creosote-pitch firing of *A* Furnace, the average heat input is 4,300,000 B.Th.U./ton of steel, but for one particular week with favourable conditions, the heat-input figure was as low as 3,700,000 B.Th.U./ton of steel. About 75–80% of the charge consisted of hot metal.

SOAKING PITS

Submitted by the Society of Furnace Builders

FOREWORD

By H. Southern, Chairman of the Society of Furnace Builders

This paper has been prepared by a Committee of the Society of Furnace Builders from information supplied by members of that organization. The main features and operation of four types of soaking pits of modern construction are described, without reference to others which are available.

During the compilation of the paper consideration was given to the inclusion of data upon the design and operation of furnaces which cannot be classified as soaking pits, but which are employed for ingot-heating purposes in certain mills where the conditions of application are suitable. As is well known, continuous furnaces of the fixed-hearth type have long been used for the heating of a limited range of ingots to rolling temperature, and continuous furnaces of the bogie type are now being employed for a wider range of ingots and slabs. Detailed reference to these furnaces is omitted from the present contribution.

The pits reviewed in this paper are as follows :

(1) Reversing pit of the conventional type with new features.
(2) Circular tangentially fired pit of the recuperative type.
(3) One-way-fired pit of the recuperative type.
(4) Bottom-fired-pit of the recuperative type.

These four types have been selected because the methods employed for flame propagation are different, the conditions of gas flow are dissimilar, and the heating chambers contrast in size and shape ; yet all successfully perform the process for which they were developed. One is equipped with regenerators and a metallic recuperator, one has metallic recuperators throughout, and two employ refractory recuperators of different type.

John Gjers of Middlesbrough introduced the soaking pit and announced his invention at the Vienna Meeting of The Iron and Steel Institute in 1882. He believed that its adoption would eliminate the use of fuel between the blast-furnace and the finished rolled product and, under the conditions of his time, he completely achieved his aim. The construction of the simple unfired pits originally installed at Darlington is well known, and they dealt with 10-cwt. ingots at the rate of 300 per shift. The inventor estimated that 2 cwt. of coal were saved per ton of rolled product and he valued this at 8*d*. He thought that savings in other directions amounted to 10*d*. but he believed that the overall savings in some works would be 3*s*.—a delightful comment evincing the pride of Gjers in the efficiency of the plant with which he was associated. He subsequently found that the scaling losses in the pits were much less than in the single and double heating processes which they eliminated and he believed that the gain in finished steel was 2%.

Sixty-five years have elapsed since Gjers spoke in Vienna and in that time the ever-increasing demand for steel created conditions which necessitated a return to the use of fuel between the melting shop and the mill. Thus the simple coal-fired soaking pit came into being and with it the producer-gas-fired regenerative reversing pit. Then followed exacting demands for greater pit capacity, improved heating conditions, reduced scaling losses, increased thermal efficiency, lower operating costs, and the wider use of low-grade fuels. Each factor had its effect upon the design and construction of pits, and by a gradual process of evolution the installations described in this paper have been independently developed. Although they differ in principle and construction the aims of the designers have been as follows :

(1) To minimize the consumption of fuel between the casting bay and the first pass of the mill.
(2) To provide thermal conditions which ensure the uniform heating of ingots of diverse size and shape at a regular rate of production without impairment of the steel.
(3) To produce and easily maintain atmospheric conditions which are not conducive to the excessive formation of scale.
(4) To reduce the costs of operation by adequate mechanization.
(5) To improve and maintain consistent conditions of combustion and heating by the use of control instruments.
(6) To reduce maintenance and upkeep costs.
(7) To provide working conditions which are less trying to the operatives.

REVERSING-TYPE SOAKING PIT

In the development and design of soaking pits the attainment of uniformity in the temperature of the heated ingots has to be duly considered in relation to the achievement of thermal efficiency.

The steel must be delivered to the mill in the most suitable condition for rolling, and whilst the economical use of fuel is one essential, other important factors have to be considered. These are summarized below :

(1) Oxidation losses should be a minimum.

(2) The pit should be capable of heating ingots at a rate which will ensure regular and continuous production at the mill.

(3) The degree of uniformity in heating should be such that fluctuating power demands at the mill motors are minimized.

(4) The conditions of heating should ensure the constant production of sections and plates of the required quality and finish.

With regard to (1), a coating of scale is required to protect the steel from damage by burning or washing, but it should be as light and friable as possible to ensure removal at the first pass in the mill. Items (2) and (3) directly depend upon adequate

Manuscript received 15th December, 1947.

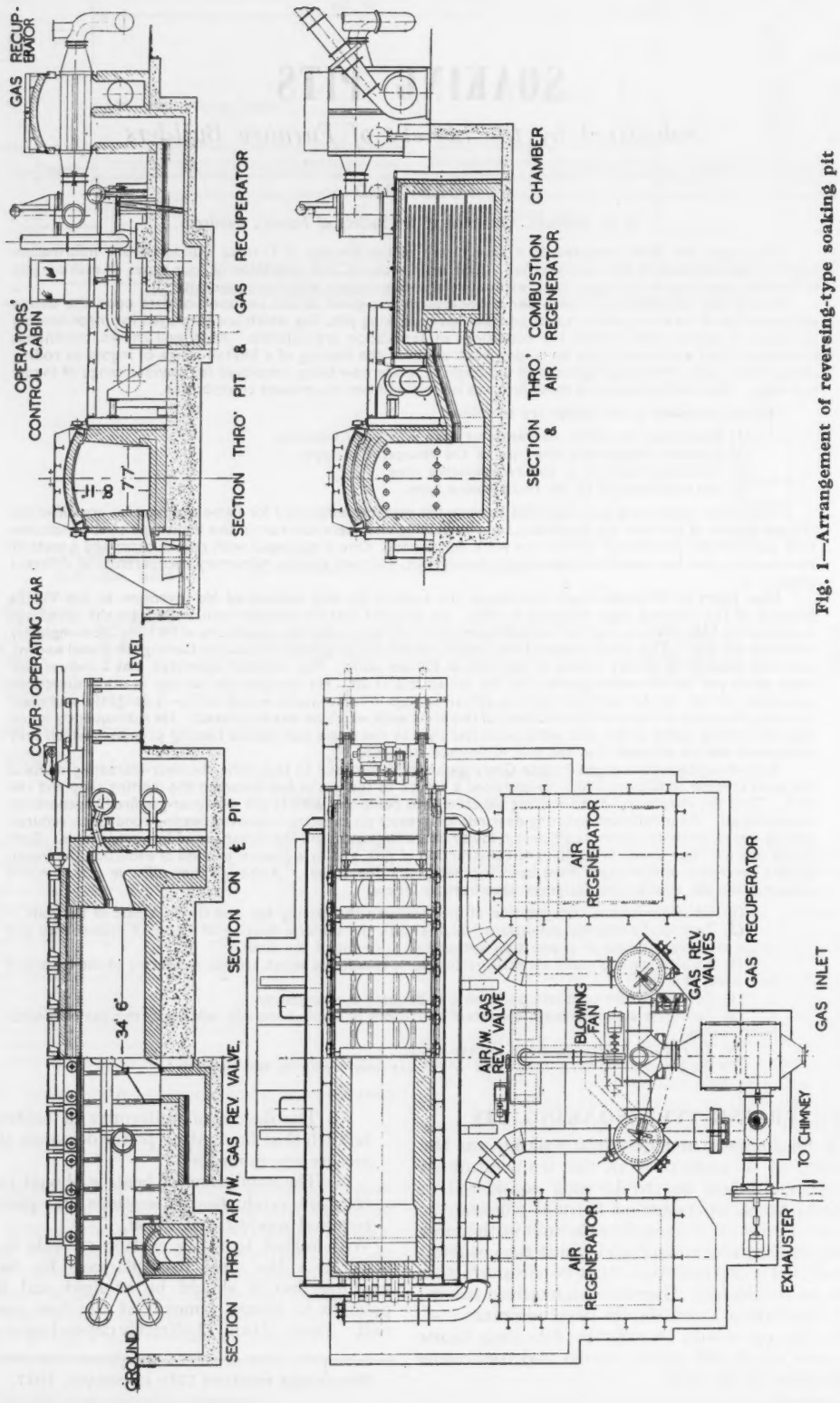


Fig. 1—Arrangement of reversing-type soaking pit

pit capacity, combined with good heating practice. In regard to (4), it is well known that defects in the semi-finished or finished product can arise from factors in the various stages of heating to which ingots, blooms, or slabs are subjected. In the particular case of soaking pits, the period of time between the casting of an ingot and its charging into the pit has to be related to the heating and soaking periods in the pit to ensure the attainment of the essential uniformity of temperature in the ingot structure.

It is important to realize the significance of the rate of heating, which must be gradual to ensure continuous and steady thermal absorption, thus achieving temperature uniformity in the ingot without damage at the surface. The use of automatic control equipment contributes in no small measure to the attainment of the essential conditions of heating already outlined. Such apparatus is widely employed today to maintain the correct air–gas ratio in relation to the temperature and pressure conditions within the heating chamber of the pit.

Reversing-Type Soaking Pit

A description is now given of a conventional type of soaking pit of modern design, two of which have been installed during recent months. Both are now in regular operation. The arrangement of each pit is shown in Fig. 1 and both are illustrated in Fig. 2.

The heating chamber is of rectangular form and has the following dimensions at hearth level:

Length	34 ft. 6 in.
Width	7 ft. 7 in.
Height to soffit of cover arch ...	8 ft. 11 in.

The number of ingots normally charged is 20, each weighing 6 tons, or a similar number of 4-ton ingots.

These pits are of the air-regenerative and gas-recuperative type, this combination of systems being extremely effective in the maintenance of high flame temperatures from lean gases.

Fuel and Firing

The fuel at present employed is cleaned blast-furnace gas with a calorific value of 100 B.Th.U./cu. ft. It is delivered to the recuperator inlet at a pressure of 5 in. water-gauge.

At each end of the pit there are two banks of burners giving a total firing range of ten gas inlets. The lower set of burners may be regarded as air–gas pre-mixers located in a primary combustion zone. The upper set are directional or boosting burners. Generous space is provided for efficient combustion with maximum thermal transfer and uniform heating of the ingots to a mean temperature of 1150° C. is attained.

Complete operating data with heat balance are not yet available for this installation although highly satisfactory performance is reported. The pit was, however, designed to give the following results:

(*a*) Charged with twenty 4-ton cold ingots; pit standing at 1350° C.; heating to rolling temperature in 8 hr. on a gas consumption of approximately 16,000 cu. ft./ton.

(*b*) Charged with twenty 6-ton cold ingots; pit standing at 1350° C.; heating to rolling temperature in 10 hr. on a gas consumption of approximately 16,500 cu. ft./ton.

Fig. 2—Battery of reversing-type soaking pits

(*c*) Charged with twenty 4-ton ingots at 900° C.; pit standing at 1350° C.; heating to rolling temperature in 4 hr. on a fuel consumption of approximately 8500 cu. ft./ton.

(*d*) Charged with twenty 6-ton ingots at 900° C.; pit standing at 1350° C.; heating to rolling temperature in 5 hr. on a fuel consumption of approximately 7000 cu. ft./ton.

Gas Recuperator

The recuperator is of the Newton metallic needle type, consisting of 50 elements arranged in a single group, 10 wide by 5 high. The elements are of heat-resisting iron and a centre baffle plate is incorporated to provide two-pass flow on the waste-gas side. To ensure access for cleaning or inspection, the refractory-lined hood is so arranged that it can be lifted clear.

The recuperator was designed for the following duty:

Quantity of waste gases ...	440,000 cu. ft./hr.. N.T.P.
Temp. of waste gases entering	550° C.
Temp. of waste gases leaving...	405° C.
Quantity of fuel gas	190,000 cu. ft./hr.. N.T.P.
Temp. of fuel gas entering ...	45° C.
Temp. of fuel gas leaving ...	375° C.
Draught loss, waste-gas side ...	1 in. water-gauge
Pressure loss, fuel-gas side ...	0·9 in. water-gauge

Air Regenerators

The air regenerators are of the horizontal multi-pass type, the approximate dimensions of the checkers being 14 ft. 0 in. long, 11 ft. 0 in. wide, and 10 ft. 6 in. high. There are six passes, with 4½-in. square apertures arranged four in depth at the first pass, and three in depth at each of the other five passes. The average brick thickness is 3 in. The generators were designed for the following duty:

Waste-gas inlet temperature (estimated)	1250° C.
Waste-gas outlet temperature ...	710° C.
Total waste-gas travel (approx.)	72 ft. 0 in. mean

The air/waste-gas reversing valve is of the orthodox type fitted with electrically operated reversing gear. The gas reversing valves are of the water-sealed type with motorized reversing mechanisms.

Control Equipment

The control equipment consists of the following items :

(1) *Auto Reversing Unit*—This unit is capable of giving 10–30-min. intervals with overriding push-button operation to ensure reversal at any time. The following is the sequence of events on reversal : Gas control valve closes ; gas and air/waste-gas reversing valves operate ; interval of 5–20 sec. to purge products from air regenerators ; gas control valve opens.

(2) *Remote Manual Control*—Lockheed cylinders are fitted to the gas, air-exhaust, and dilution control valves to give manual control from a central position.

(3) *Auto Combustion Control*—A Reavell Askanita apparatus ensures a constant air–gas ratio ; an orifice plate is fitted on the air and gas mains, with an auto-controlled valve in the air mains.

(4) *Pressure Control*—A Reavell Askanita apparatus maintains constant pressure in the furnace chamber. There is a pressure-tapping in the centre of the side wall ; the exhaust valve is operated through the instrument in ducting between the exhauster and the chimney.

Solenoids are fitted to gas and exhaust valves to ensure the following sequence : Gas-valve closure before reversal ; gas-valve closure in event of failure of power to air fan ; gas and exhaust valves closed before opening of pit covers.

Temperature control is manual from an indicating pyrometer.

Features of Construction

The thickness of brickwork in the hearth is 1 ft. 11 in., made up of 9 in. of "Thistle," 4½ in. of "Nettle," and 9 in. of magnesite paving. Chrome-magnesite bricks are employed for a certain distance up the side walls and also around the slag notches. The mean thickness of the side walls is 2 ft. 0 in. The body of the pit, the regenerators, and the recuperator are adequately insulated.

The framework forming the rolling covers is of particularly robust design in welded construction, the platework varying from 1 to 1½ in. in thickness. The

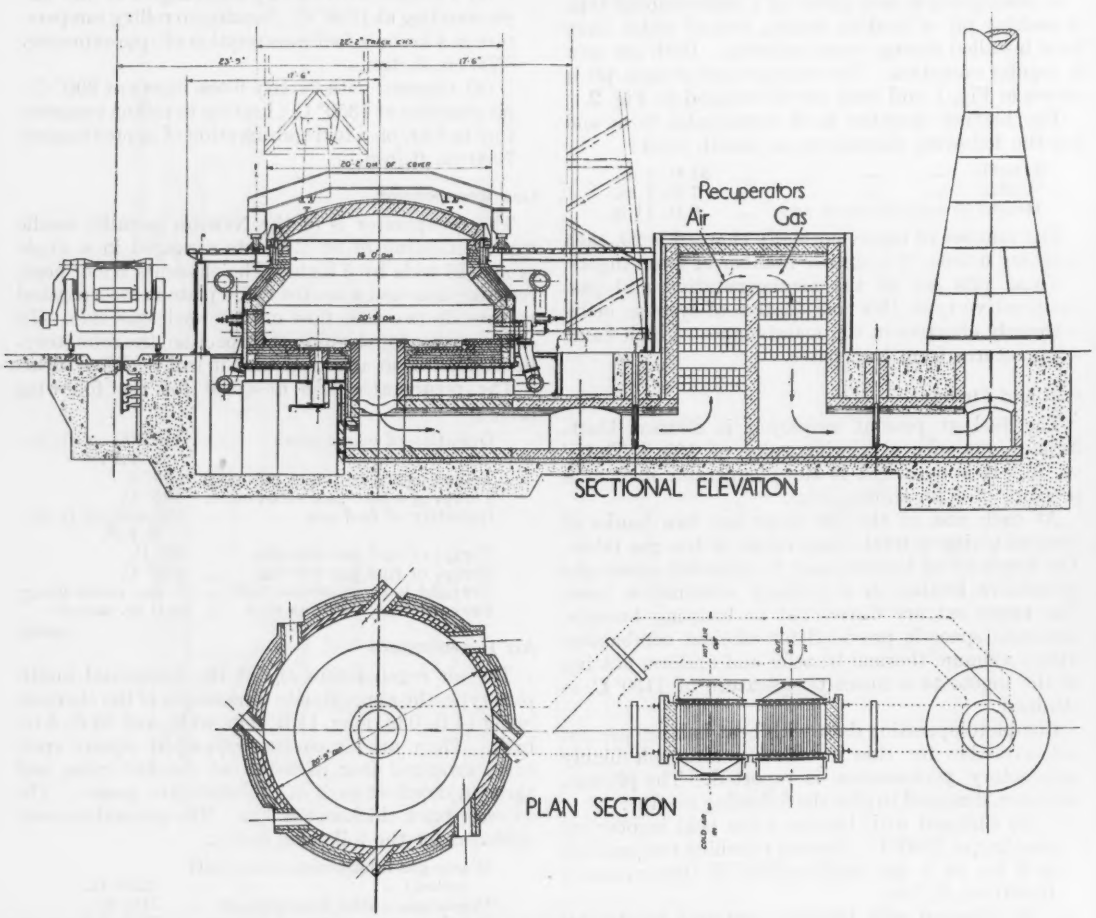


Fig. 3—Arrangement of circular soaking pit

cover-operating mechanism is electrically driven through machine-cut reducing gears to pin-rack-type rams, one at each side to maintain alignment. Special study was given to the design of the covers to ensure a thoroughly sealed job. The extent of the lap between the end cover and the ramp is about 3 ft. 6 in., thus preventing any serious gas escape at this point.

Although there are many types of pit cover, and all have merit, the split rolling cover has certain advantages. It is unnecessary to expose a large area of the pit when gripping and lifting an ingot; there is no restriction of the view from the crane cabin; and neither the machine nor the operator are subjected to trying conditions.

The body of the pit is totally encased in steel plates with heavy supporting buckstays and binders. Complete encasement is always essential to prevent air infiltration or gas leakage and is vitally necessary when clean gases are employed.

Insulation was adopted for the dual purpose of conserving heat and of sealing the pit, the regenerators, and the recuperator. The various refractory materials were carefully selected and considerable care was exercised in the provision and location of suitable clearances for expansion.

The pit has several features of unusual interest and is a flexible unit under normal conditions of operation. The achievement of optimum operating efficiency nevertheless depends upon:

(*a*) Regular continuity of loading under experienced supervision.

(*b*) Proper care and use of the control instruments.

(*c*) Maintenance of the pits in sound condition.

CIRCULAR SOAKING PIT

The design of this furnace radically departs from that of the conventional type of soaking pit in that it is circular in form, and equipped with a multiplicity of burners which fire into the heating chamber tangentially, the waste gases being evacuated by means of a central port in the furnace base. The gases circulate around a combustion chamber formed by the sloping walls behind the ingots, work their way through the ingots and to the top of the pit, are then pulled downward to the flue exit in the centre of the pit bottom, and thence to a recuperator or direct to the stack, depending on the particular installation. Figure 3 shows the arrangement of a circular soaking pit and a battery of such pits is shown in Fig. 4.

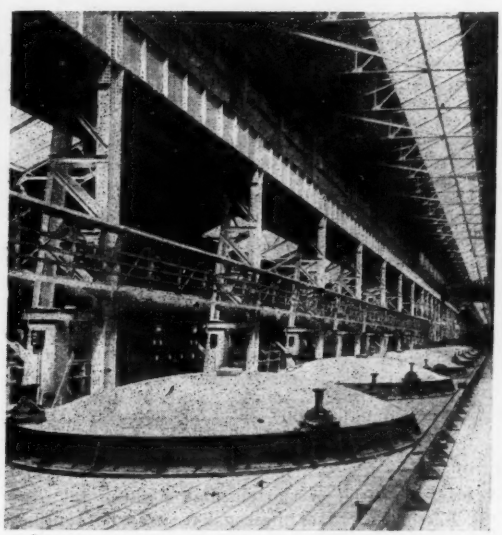


Fig. 4—Battery of circular soaking pits

Several different sizes of circular pits have been built, ranging from 11 to 20 ft. in dia. The most practical of the various sizes are the smaller pits of 11–16 ft. in dia., and the choice is dependent on the sizes of ingots to be rolled and the capacities of the furnaces producing the ingots. Any type of fuel can be used and installations are now operating on oil, natural gas, mixed coke-oven and blast-furnace gas, hot raw producer gas, and clean blast-furnace gas.

Method of Firing

The burners are set at an angle to the periphery of the furnace, and the fact that the actual heating gases are moving in a spiral path makes a high rate of circulation possible with relatively low burner velocities. There cannot be on this design of soaking pit any stagnant pockets of gases. It is estimated that the whirling action of the gases within the circular pit moves at a speed of 50 ft./sec. and make at least two complete revolutions in the furnace before entering the waste-gas flue. The gases thus travel approximately 100 ft. or more, which is from three to ten times as far as in any other type of ingot-heating furnace.

The burner ports are relatively long so that combustion is substantially completed in them. In addition the fresh products of combustion from each burner are mixed with the volume of waste gases circulating in the pit which have already given up heat to the ingots, thus tempering the total heat head or differential in the combustion chamber behind the ingots.

Control

Completely automatic control is a feature with the circular pit, as it is in all modern types, so that operation is not left entirely to the heater.

Calibrated instruments show exactly what is going on within the pit during the entire course of the heating cycle; the instruments controlling pit temperature, the exact ratio of fuel to air, the pressure in the pit, and the rate at which the fuel is being burned. Consequently, having decided on the best heating cycle for any particular class of steel, the instruments are set to the desired settings and these conditions can be reproduced day after day giving correctly heated ingots free from washing and with controlled scale jacket. Naturally, all instruments require regular maintenance service.

Owing to the uniformity of temperature in the circular pit which does not vary more than $\pm 10°$ C., only one control thermocouple is necessary and this is set just above one of the burners, or alternatively in the waste-gas flue outlet. A considerable amount

of the uniformity of heating is due to convection as well as radiation because of the high velocity of the gases surrounding the ingots.

Features of Construction

Pit Bottoms—In most circular pits coke breeze is used for making bottom and the average consumption of breeze is of the order of 10 lb. per ton of steel heated. It is obvious that only coke with a high ash-fusion point should be used. Both in America and in Sweden, however, circular pits are in operation working with a dry bottom. On such installations there is only a very small build-up of slag over a lengthy period; this slag is dug out when the pit is shut down for repairs.

Cover and Cover Crane—The furnace cover, in common with all other modern types of pits, is lifted vertically from the sand seal and then traversed laterally to open the pit to the desired extent. It is necessary only to withdraw the cover a sufficient amount to charge or discharge any particular ingot. The sand seal effectually prevents the flame from blowing out round the base of the cover, the former being a trough located on the outside of the furnace and so cooled by exposure to air. The furnace cover is provided with heat-resisting alloy dippers which give an efficient seal. The cover crane has two motors, one for lifting the cover, and a smaller separate motor for travelling. During normal operation the crane is moved into position over the top of the desired furnace, and whilst in this position limit switches operate which permit the crane to raise the cover clear of the sand seal. When the cover is raised to its full height the limit switches are automatically reset and it is then possible to move the crane and cover to either side of the pit so that ingots can be charged or drawn. It is impossible when the crane is not in the correct position over the pit for the operator to lower the cover.

Fuels and Recuperation

The circular pit operates very successfully when fired with hot raw producer gas and cold air, although naturally at the expense of fuel and in the interests of fuel conservation recuperation of the combustion air is recommended. As, however, recuperators from time to time require overhauling and possibly some repair, the fact that a pit fired with hot raw producer gas can still operate satisfactorily with cold air whilst repairs to the recuperator are carried out obviates the necessity of shutting down the pit with consequent loss of production. Facilities are provided for by-passing the recuperator, allowing the waste gases to go direct to the stack should repairs be necessary to the recuperator.

When the richer fuels are adopted for firing, recuperation of the air is not necessary to obtain flame temperature and in such instances waste-heat boilers may be installed with resultant fuel economy.

Fuel consumption per ton of ingots depends upon many factors such as temperature of ingots when charged, *i.e.*, track times, heating rate desired, soaking period, etc., and whether recuperation is installed, and the circular pit compares favourably with any other modern pit operating under exactly similar conditions.

Turning to the question of firing with blast-furnace gas, it is necessary when this fuel is used for both the gas and the air to be preheated, and metallic recuperators of the needle or tubular type are usually used.

With this type of recuperator, however, it is necessary to limit the temperature of the waste gases entering the recuperator to a maximum of 1050° C., and dilution of the waste gases by means of cold air is provided for the purpose. It is of interest to know that comparatively recently a metallic recuperator in which the initial elements are protected with refractory sheaths has been installed on a circular pit with the object of eliminating the dilution air; operation to date has shown that this method will ultimately be quite successful, although there still remain a few small practical details to improve.

Operating Data

Actual operating data are always interesting, and the following figures have been taken from 16-ft. dia. circular pits in recent years, operating under the different conditions as stated.

(a) *Hot Raw Producer Gas with Air Recuperation*

Average weekly tonnage per pit...	1300 tons
Output per hour, cold steel ...	6·6 tons
Output per hour, black hot steel	8·6 tons
Output per hour, hot steel ...	13 tons
Average fuel consumption ...	1¼ cwt. per ton of ingots

(b) *Mixed Blast-Furnace and Coke-Oven Gas without Recuperators*

Average weekly tonnage, all pits	10,060 tons
Average weekly tonnage per pit (30% cold steel) ...	1437 tons
Maximum tonnage per pit per week (10% cold steel)	2140 tons
Fuel consumption with no recuperators	1·3 million B.Th.U./ton (average)

(c) *Blast-Furnace Gas with Gas and Air Recuperators*

Tonnage per week, all pits...	18,000 tons
Average per pit per week ...	2250 tons
Cold steel 5%; hot steel 95% with 1hr. 40 min. track time	95%
Calorific value of gas ...	103 B.Th.U./cu. ft.
Fuel per ton of cold steel ...	15,530 cu. ft.
Fuel per ton of hot steel ...	7060 cu. ft.
Average fuel over 3 months...	9000 cu. ft.
Temperature of air	600° C.
Temperature of gas ...	400° C.
Duration of heating time ...	Cold ingots, 8 hr.
Duration of heating time ...	Hot ingots, 3 hr.

At an official acceptance test made at the above plant the following figures were obtained:

Cold Steel:	
Heating time	8 hr. 12 min.
Fuel consumption ...	1,405,000 B.Th.U./ton
Scale loss	0·58%
Hot Steel:	
Heating time	2 hr. 25 min.
Fuel consumption ...	538,000 B.Th.U./ton
Scale loss	0·53%

Summary

Special features of the circular soaking pit are:

(1) Accurate control of air–gas ratio and of temperature makes it possible to hold ingots for several hours when necessary because of mill delays, etc., without

any deleterious effect on the steel and with only a slight increase in the thickness of the scale jacket.

(2) Fuel being supplied to the furnace at a number of points ensures that no section of the pit is at any time receiving a large mass of flame, and consequently individual ingots do not tend to get overheated.

(3) With tangential firing not only is there no direct flame impingement on the ingots but also there is absolute temperature uniformity throughout the furnace. Actually the heat supplied, particularly towards the end of a heat, is purely radiation from the surrounding brickwork.

(4) With such uniformity of heating, temperature control becomes extremely accurate and it is possible for ingots to be drawn from the pit varying not more than 20° C. from top to bottom.

(5) Owing to the fuel being distributed at a number of points, faster rates of heating, particularly with steel charged cold, can be safely employed without the danger of burning any of the ingots, and provided that the temperature control is not set too high and the air–gas ratio is properly adjusted, it is impossible to wash an ingot.

(6) The circular construction gives a strong design and in particular, having a continuous outer steel casing, expansion of brickwork can be controlled by means of expansion joints in the casing.

ONE-WAY-FIRED SOAKING PIT

An old textbook, "Metallurgy of Iron" (1890), states "Besides the saving in fuel, the soaking pit has the important advantage of maintaining a neutral or reducing atmosphere about the ingot, so that the surface oxidation is less than in an ordinary heating furnace." This section of the paper deals specifically with the design of the one-way-fired pit and sets out to prove the efficacy of this type for high-quality and economical heating of ingots. The principal feature of the design is the combustion of the fuel above and the descent of the heating gases around and enveloping the ingots (vertical horse-shoe firing). Thus the heating is carried out in the best possible atmosphere for quality product.

Thirty years ago one-way-fired pits in batteries of several pits, but each holding one ingot only, were installed in France.

Twenty years ago the first pits of this type were built in Britain, but with an important difference —each pit held six ingots of about 50–60 cwt. each. Air only was preheated in recuperators and, even so, cold blast-furnace gas was successfully used, though coke-oven gas is the more usual fuel. The pits were built in one battery of six.

Associates in the U.S.A. were also developing this design, and with their greater opportunities in capacity were proceeding at a greater pace. The first British pits mentioned above were only 9 ft. long, but in the U.S.A. they were being built 12 ft. and 14 ft. long.

It soon became apparent that the length could be increased, and by careful attention to burner design and flame propagation along with the general proportions of the pits, the next stage was a number of installations ranging from 16 ft. to 20 ft. long; within the last few years pits 23–26 ft. long have been built.

It will be appreciated that, except for differences occasioned by these remarkable increases in length, the basic design remains the same, thus emphasizing correctness of its principle and of its natural flame development and flow.

Each individual pit of the battery of one-way-fired pits (as shown in Figs. 5 and 6) is 25 ft. long by 7 ft. 6 in. wide. The floor space occupied by each battery of pits is very much less than with any other design. The pit proper is in the soaking-pit building, the back wall of the pit being arranged at about a minimum approach of the ingot grips; there is ample space between the front of the pits for an ingot chariot and two or more ingot roads. Pits of this size will hold 100–120 tons of ingots, and it will immediately be appreciated that the above claim regarding floor area occupied can be readily substantiated with ingot capacities of this order. With a layout as shown in Fig. 6, repairs to the lids are easily carried out in the lean-to and valuable crane space in the soaking-pit bay is not occupied for such work.

It is well known that the time required to bring the ingots to rolling temperature after charging in the pits is a function of their charged temperature, and experience has shown that this is closely related to track time, *i.e.*, the time between teeming and charging the ingots into the pit. For instance, if the time between teeming and charging is 3 hr., the approximate heating time will be 3 hr. + 1 hr. The

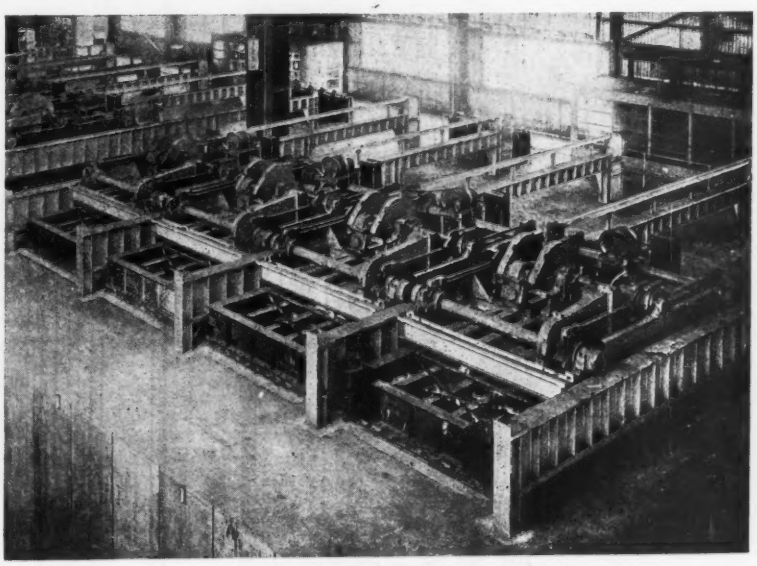


Fig. 5— Battery of one-way-fired soaking pits

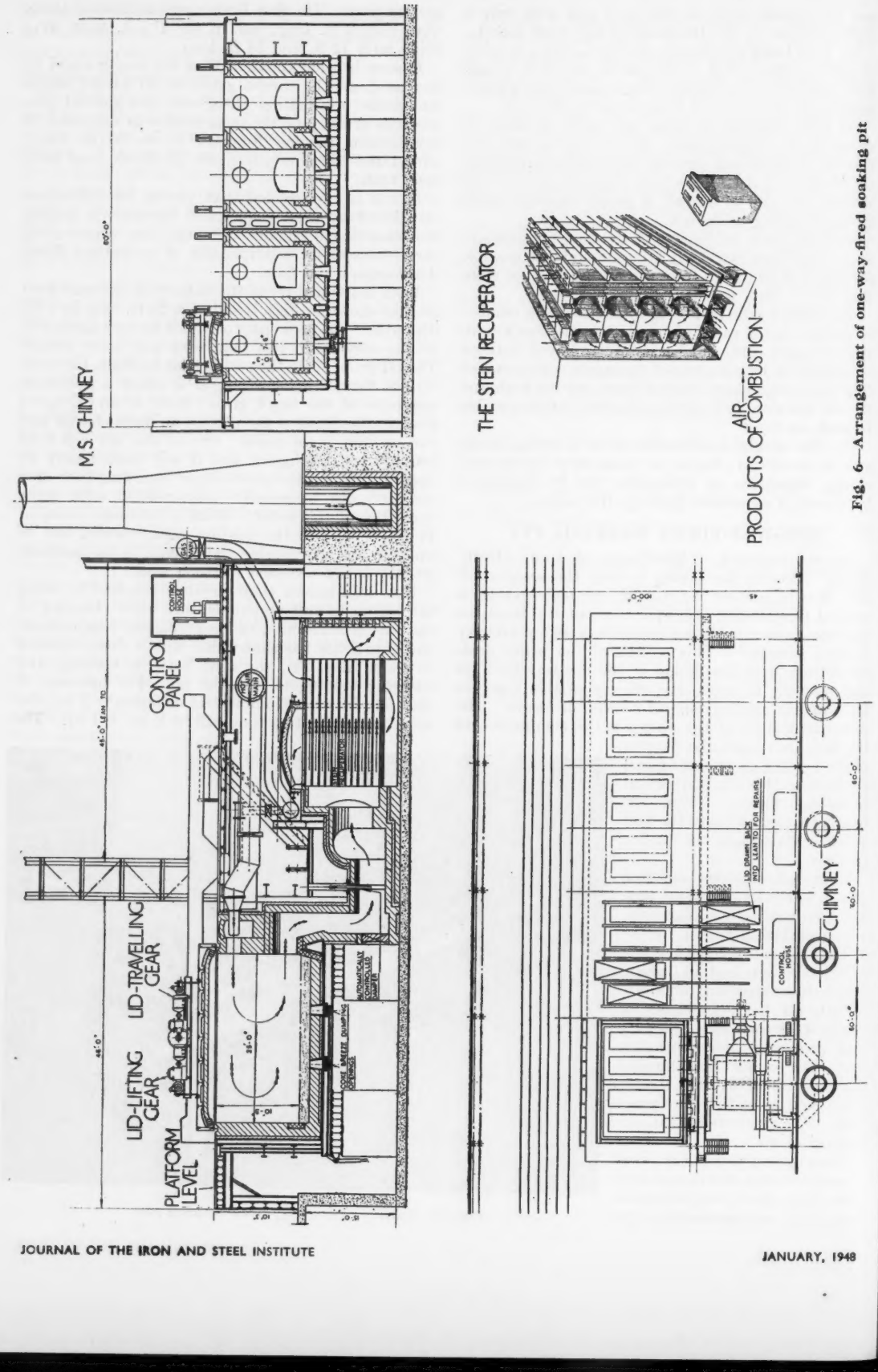


Fig. 6—Arrangement of one-way-fired soaking pit

times for heating cold ingots depend upon the thickness of the ingot, and a rough guide is about 22 min. per inch of thickness. Large rectangular slab ingots take a little longer owing to their mass. The heating of ingots should not be forced. It is far better to have slower heating in the correct atmosphere than flash heating, with washing of the surface of the steel, the production of large quantities of slag, and higher maintenance of the pit brickwork consequential upon forcing the pace. In planning any installation of pits, ample or excess capacity should be provided. This is, of course, more readily possible with the one-way-fired pit, owing to the smaller area occupied.

As mentioned earlier, whilst the one-way-fired principle has been maintained throughout the development of the pits, improvement in detail has been progressive, the greatest development being in the separation of the air-preheating system from the pits proper; the air recuperators are now built at the rear of the pits, one recuperator being common to a number of pits instead of one recuperator per pit as originally. This arrangement has had other advantages in that it enables the benefits of fully automatic control of temperature, combustion, and pit pressure, to be obtained in a simple and positive manner. Further, slag cannot reach the recuperators and the carry-over of dust is minimized, this being largely collected in the pockets under the pits and the flues connecting these with the recuperators. The life of such recuperators is vastly extended.

Fig. 6—Arrangement of one-way-fired soaking pit

Oxidation Losses

Minimum loss by oxidation is of the highest importance, for a difference of only ½% can represent many thousands of pounds per year in saving of steel, i.e., increased yield. Furthermore, scale or slag costs money to remove from the pits and from the mill. The one-way-fired pit with its extensive combustion space above the ingots prevents stratified flame contact with the ingots, these being enveloped only by fully combusted gases.

Oxidation is a function, not only of pit atmosphere, but also of time, and obviously the loss is greater on cold ingots. By reason of the difficulties of making practical tests, definite investigation of relative loss with hot and cold ingots has received scanty attention. Tests in Great Britain on one-way-fired pits built twelve years ago, when heating cold ingots weighed before charging and with the resultant blooms and crops weighed after rolling, showed a loss of 1·29%, the pits being hand-controlled in their normal manner of working along with many other pits and without special supervision. Such loss is appropriately less on hot ingots, whilst the relative amount of scale left in the pits or brought out with the ingots is again a function of time and atmosphere. The use of combustion and temperature controls is of prime importance in successfully reducing oxidation loss and will be referred to later.

Fuel and Firing

In regard to the firing system, the provision of a single burner in the back wall of the pit makes it possible for this to be of the correct design for proper combustion and flame propagation over the ingots. There is no compromise as far as the design of the burner is concerned, as is the case with regenerative pits; it has only one purpose and is not at one time a burner and another a waste-gas port, neither can it be damaged by slag or ingots. The burner can be made suitable for gas or oil, or a combination of both. Blast-furnace gas, raw producer gas, coke oven gas, or a mixture of blast-furnace and coke-oven gas can all be used with an easy change-over. In the case of oil, with which these pits are extensively heated, the change is extremely simple. If blast-furnace gas alone is available, then preheating of this is desirable in a metallic recuperator. For mixed blast-furnace and coke-oven gas firing, a lower calorific value of around 130 B.Th.U. is used with preheating of the combustion air only. The burners are of special design, and the rate of mixing is varied according to the type of fuel being used in order to give a long flame travel, but with low directional speed and without harshness even with oil.

Minimum fuel consumption is achieved. Weekly figures show 1,800,000 B.Th.U./ton on cold ingots and 650,000 to 800,000 B.Th.U./ton on hot ingots. Figures for individual heats show as low as 1,375,000 B.Th.U./ton on cold ingots and 400,000 to 500,000 B.Th.U./ton on normal hot ingots. With raw producer gas and on hand control in a large mill in Great Britain, the fuel is only 1·15 cwt. of coal per ton with 53% hot, 16% black hot, and 31% cold ingots.

Recuperators

With regard to the recuperators, the most widely used for these pits is the Stein refractory type, in which the waste gases pass horizontally through passages formed between the air-heating blocks, whilst the air passes vertically through the holes in the blocks. In order to keep a balance of pressure between waste gases and air, and hence to minimize the permeability of a refractory structure, and to facilitate combustion control of the pits, a hot-air fan is used to draw the air through the recuperators. With the materials at present available for economical manufacture of the fan, the temperature of the preheated air is limited to 1100° F. (590° C.), which is fully satisfactory for fuel economy and efficient flame generation. With the recuperators built away from the pits, and with the preheated air handled by a fan in the manner described, the life of the recuperators is very many years.

Slagging

With regard to slagging, the development in the U.S.A. has principally led to the use of coke breeze with proper facilities for the handling of the breeze and the waste material. In this country, although considerable interest is being shown in the coke-breeze bottom, it is more usual to allow the slag to build up in the bottom of the pit and to shut down the pit periodically and dig out the slag. With this practice the pits have a run of some 8–10 weeks before the slag has to be removed.

The normal arrangement with coke-breeze bottoms is to have two openings through the bottom of the pit for the discharge of the slag-impregnated coke,

which is broken up by a rake held by the ingot grips, the spent coke being raked towards the openings and discharged into trucks in the basement. Fresh charges of coke are dropped into the pit from bottom-opening hoppers, also carried by the crane, the coke being tamped in place by a ram held by the ingot grips. The consumption of coke, which must have a high ash-fusion temperature, is about 1¼ tons per week on a 25-ft. pit. The operation of bottom-making takes ½–1 hr. and has to be performed every 7–12 days. A complete breeze bottom lasts for 4–5 months.

Lids and Lifting Mechanisms

In the development of the pits particular attention has been paid to the design of the lids and their mechanisms. In order that the pit may be operated at a pressure slightly above atmosphere, it is necessary to seal completely the lid when it is in its final position over the pit, and it is a feature of the one-way-fired pit that a complete seal is always provided. The most usual form of lid-operating gear consists of a two-motor carriage running on tracks on either side of the pit, from which the lid proper is suspended by readily detachable means. One motor is used for the lift and lower motion, and the other for long travelling the lid. Current is brought to the lids by a flexible electric cable with suitable take-up gear, or in some cases by collectors. The firebrick crowns of the lids, which are effectively insulated, are generally of the arched type whereby a very tight construction is obtained. Suspended-type crowns are used as well, but these have generally found more favour in the United States than in this country. For rebricking, the lids are pulled back on to the platform at the rear of the pits where they can be detached from the motor carriage to give easy access for relining. The long travel of the lids on the tracks is usually arranged so that in addition to being pulled back for crane access to the ingots they can be moved in a forward direction; thus the heater can readily see into the pits.

An important development is the operation of the pit lids from the ingot crane. When this is opposite the pit the crane man, by a controller in his cab, opens the lid to the exact extent he requires. After charging or drawing an ingot, he puts the controller over and moves off. He does not wait for the lid to close; this is done electrically. This system is time-saving, heat-saving, and labour-saving, as will be appreciated from the fact that the total labour in one battery of 24 pits, each 25 ft. long, and with an exceptional variety of ingot sizes and qualities, is one heater, one bottom-maker, two crane men, one recorder, and one assistant. The heater is, of course, able to operate the lid from the pit platform. The fuel is turned off and on when the lid is opened and closed.

Control

As mentioned earlier, the pits are provided with fully automatic control of temperature, combustion, and pit pressure. The advantages of these cannot be too highly stressed. The reduction in scale, better steel yield, less slag deposit in the pits, and reduced fuel consumption effect a most substantial yearly saving. Maintenance of pit and lid brickwork and pit bottoms is reduced; definite schedules of heating can be laid down and adhered to, so that the best results for varying steel qualities, ingot sizes, and product can be obtained. The instruments are usually placed in a glass-fronted house at the rear of the platform, where they are free from dust and consequently more easily maintained in good order. In addition, instruments are usually provided to meter the flow of fuel and to safeguard the fan.

To summarize, the advantages of the one-way-fired pit are so widely appreciated that there are 120 of these in Great Britain, 295 in the U.S.A., and some 50 on the Continent.

BOTTOM-FIRED RECUPERATIVE SOAKING PIT

It will be appreciated by steelworks engineers that in any unit of their plant, the overriding consideration is efficiency, including the vital factors of power consumption, and operating and maintenance costs, and, particularly in furnaces, every attempt must be made to save fuel.

With the bottom-fired recuperative type of soaking pit, shown in Figs. 7 and 8, of which there are 170 in operation in the U.S.A., remarkable performances have been achieved whereby the fuel consumptions have been greatly reduced—in some cases by as much as 40% compared with regenerative-type pits.

The purpose of this section of the paper is to outline the reasons for this achievement and also to describe briefly the special features and operation of this pit. The pits are rectangular in size, varying from 10 ft. to 16 ft. The ingots stand in single row alongside the walls and are equidistant from the single firing port located in the bottom of the pit. The ingots thus form a hollow square in the centre which in effect is a primary combustion chamber. As the firing port is in the bottom, the pit depth is a minimum to suit the length of ingots and additional depth is not required for combustion. This pit is more compact and occupies less than half the floor space of other pits owing to the fact that the recuperators, which are considerably smaller than regenerators, are an integral part of the pit construction.

Fuel and Firing

All types of gases and/or oil can be used; cold 90-B.Th.U. blast-furnace gas has been used, although it is preferable to preheat this or to boost with oil. The fuel is directed vertically upwards into the pit after mixing at the bottom of the firing port with the combustion air preheated in the recuperators to 1650° F. in the case of rich fuels or to 1900° F. in the case of blast-furnace gas.

The heat distribution to all the ingots is by natural convection or fountain-like flow along the axis of the ingots and, as the gases increase in velocity immediately on combustion, turbulence is increased, causing recirculating by induction of some of the earlier-released gases which have circulated among the ingots and travelled vertically downwards towards the firing

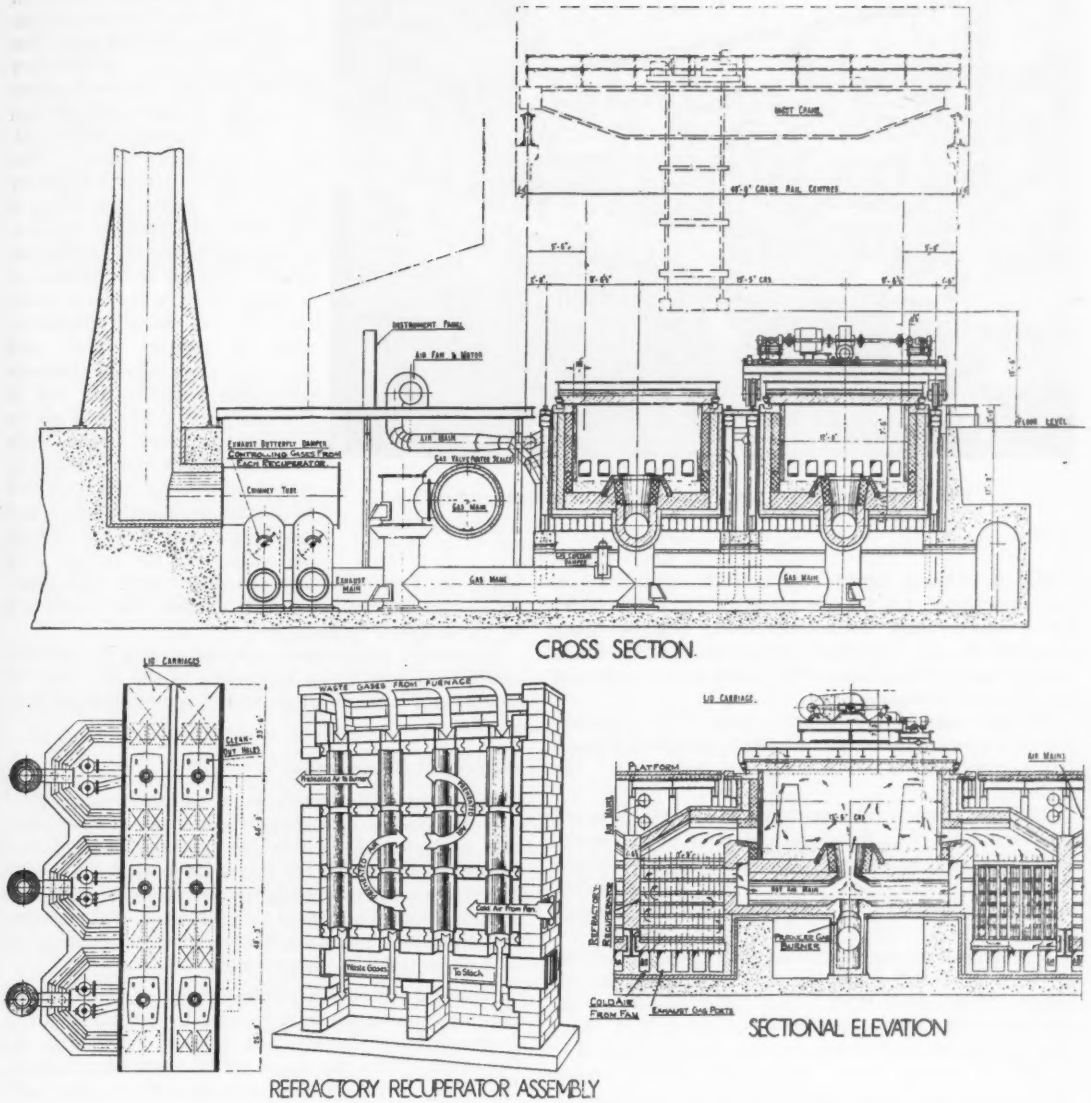


Fig. 7—Arrangement of bottom-fired soaking pit

Fig. 8—Battery of bottom-fired recuperative soaking pits

ports or exhaust ports. This method of firing is an application of the basic principle of heat transfer by radiation and convection and, besides being economical, prevents direct impingement of the flame on the ingots, thus eliminating washing of ingots and ensuring even heating.

In the case of regenerative pits or pits having burners around the walls or at the ends, the highest flame temperature is between the ingots and the walls or cover.

The net heating time required in the pit is based on track time, *i.e.*, from start of pour at the open-hearth furnace to charging into pit, plus 1 hr. when track time is between 1½ and 4 hr.; if over 4 hr., the net heating time in the pit is equal to the track time.

The overall heating rates per month for this type of pit in the U.S.A., with the usual percentage of cold ingots, is 650,000 B.Th.U./ton, and based on *all cold* ingots the heating rate is 1,600,000 B.Th.U./ton. For hot ingots the fuel rate averages 140,000 B.Th.U./ton/hr. of heating time. These very low fuel rates are attributable to the method of firing, recuperator efficiency, and complete automatic control.

Scale Loss

Because of the automatic control and continuous firing, the ingots are not subjected to temperature shocks; therefore the light scale formed is of uniform texture and falls off easily on the ingot contacting the mill table or transfer bogie. The reduction in scale compared to that formed in regenerative pits has been 40%, and on several installations the scale loss has actually been 0·8–1·0%.

Recuperator

The waste gases are removed through ports near the four corners of the side walls and at a minimum height above the coke breeze or cinder in the pit bottom. There is one recuperator at each end of the pit; this operates on the counter-flow principle with the waste gases travelling downwards through the thin-walled octagonal-shaped refractory tubes. The tubes have high conductivity and low porosity, and owing to the very thin walls give maximum heat transfer in minimum space. The air to be preheated is supplied by a fan and enters the base of each recuperator after a control valve and travels horizontally in alternate directions towards the top of the recuperator where it enters a collecting flue which leads direct to the burner. Adjacent recuperator tubes are staggered to ensure slight turbulence and maximum contact of the air with the surface of the tubes. All joints are horizontal and are sealed with high-grade air-hardening cement which ensures that there will be virtually no leakage.

Automatic pit-pressure control avoids air infiltration during loading or unloading of the pit, and the recuperator is therefore protected against temperature shocks.

Features of Construction

Pit Cover—The cover is a rigidly braced steel structure designed to support any make of suspended arch, and around the sides is a continuous alloy-steel dipper casting which fits into a deep sand-filled seal. The cover positively seals the pit when lowered into position, thus eliminating air infiltration and ensuring that the pit is maintained at a constant pressure of + 0·03 to 0·05 in. water-gauge.

Cover Carriage—An electric cover carriage, either cab- or remote-controlled, can be supplied for a single-span or double-span layout. (*See* Fig. 9.) The carriage lifts the cover before traversing in either direction, thus it is only necessary to uncover part of the pit. The hoisting and traversing movements of the cover carriage are interlocked; this prevents both movements from taking place simultaneously, although provision is made for these to be separately controlled. A further safeguard ensures that the cover can be raised or lowered only when the cover carriage is centrally located over the pit. In the closed position the cover structure is supported on brackets which form part of the solid pit steel binding.

The fuel, air, and chimney-draught valves are automatically and almost instantaneously closed when the cover is lifted out of the sand seal, and opened

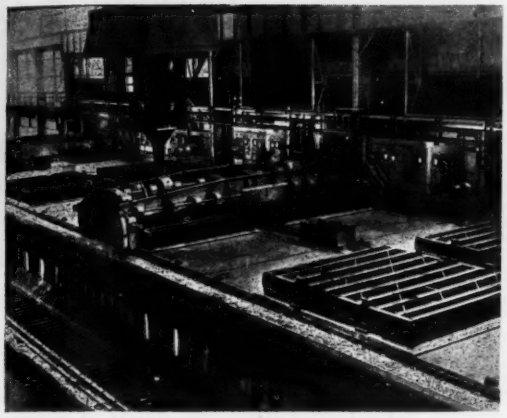


Fig. 9—As Fig. 8, showing view of cover carriage

when the cover is lowered into position in the seals. This automatic shut-down on controls, in addition to saving fuel, protects the recuperator and also the crane, making possible the speedy location and handling of the ingots.

Pit Bottom and Bottom-Making—These pits have a coke-breeze bottom, although a few are at present experimenting with a dry bottom of very clean fine mill scale.

The bottom-making period averages 10–14 days; coke-breeze additions are made at intervals of 5–6 days during continuous operation. The bottom-making operation takes 2 hr., including time required to burn in fresh coke breeze. The average quantity of breeze required is 6–7 lb./ton of steel heated.

Slag Removal—The slag and scale are removed from the pit by a spade-type scraper attached to the ingot crane which loosens the material at each of four cleaning-out doors in the pit bottom and also scrapes the remainder towards these doors, where it is discharged into small trucks which can either be run to a single hoisting station or be hoisted from the pit basement to the floor by the ingot crane.

Control

Temperature—The furnace-temperature control consists of two potentiometer temperature-controllers and one four-point potentiometer-type temperature recorder located on the panel board and noted as upper-limit control, lower-limit control, and furnace temperature. Two thermocouples are located in the furnace side wall near the top with leads running to the upper-limit controller, which is set for an upper maximum furnace temperature. Two thermocouples are located in the furnace outlets at diagonally opposed corners of the furnace with leads running to the lower-limit controller, which is set for a lower minimum furnace temperature. During the initial period of the heating cycle, the upper-limit controller functions as a furnace-temperature limit until the temperature of the waste-gas outlets reaches the set control of the lower-limit controller, at which time the lower-limit controller automatically takes over the controlling functions and continues thus throughout the balance of the heating cycle.

Combustion—The furnace combustion control consists of an air–fuel ratio regulator based on balanced differential pressures in the gas line and combustion-air line. The gas flow is measured by means of an orifice in the gas lines and the differential pressure across this orifice is imposed on the regulator which sends an impulse to the power cylinder which operates the valve in the combustion-air line to maintain this differential pressure balance over a range of flows.

The air flow is measured across an orifice in the air line ahead of the butterfly valve and the differential across this orifice is imposed upon the regulating diaphragm of the air–fuel regulator which thus balances the air flow against the gas flow and regulates the air in the ratio desired to gas flow.

Pressure—This is a standard unit continuously sampling the furnace pressure from a tube in the wall of the pit where it cannot be damaged by the ingots through lines which run directly to the regulator. The regulator controls the pilot valve which regulates the flow of oil into either side of the power cylinder, actuating a butterfly valve in the waste-gas line to the stack to maintain a predetermined pressure in the furnace.

Automatic Shut-Off on Cover Opening

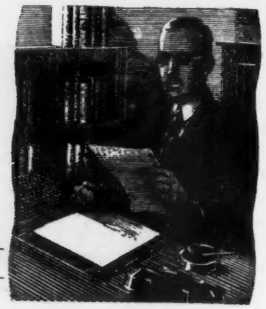
This provision is made as a practical operating feature and as a fuel-conserving feature. As the cover is hoisted the fuel is shut off, also the combustion-air-controlling butterfly valve is closed. As the protective pressure is dissipated in the furnace owing to the cover opening and the fuel and air shutting off, the furnace-pressure controller functions to close the stack draft. As the cover is moved to the closed position the gas and air flow resume and the damper opens automatically.

Industrial Radiographic Standards for Steel Castings

The American Society for Testing Materials has published two sets of standards covering Industrial Radiographic Standards for Steel Castings. Issued under the designation E 71–47 T there are radiographic negatives comprising X-ray and gamma-ray standards. Each set is priced at $30.00, or both sets (both X-ray and gamma-ray) at $55.00.

The negatives cover defects which are occasionally encountered and they are divided into groups according to the type of defect. An extensive table is used to determine the reference standard which shows the maximum acceptable amount of any type of defect. Suggestions for classifying castings to be used with the standards are given in the printed Tentative Procedure E 71–47 T. This printed procedure, giving a description of the Radiographic Standards, is available without charge from A.S.T.M. Headquarters, 1916, Race Street, Philadelphia 3, Pa., U.S.A., but, of course, it is of little service without the radiographic negatives.

CORRESPONDENCE



LETTERS TO THE EDITOR · GENERAL CORRESPONDENCE

Plastic Replicas for Surface-Finish Measurement

15th December, 1947.

THE methods hitherto described for producing plastic replicas of metal surfaces for electron microscopy or multiple-beam interferometry are not suitable for roughness measurements by stylus instruments. The reason for this is that the replicas themselves are too fragile or their production involves heating the metal specimen,* which is not always possible.

The technique outlined below, developed in the British Iron and Steel Research Association laboratories at Sketty Hall, and to be reported fully later, is of more general applicability, as it requires no heating and can be used to make negative surface replicas from objects of any size, roughness, or location.

One drop of a homogeneous liquid mixture of monomeric and polymerized methyl methacrylate, containing small proportions of a photochemical catalyst and an organic ester to act as 'stripping agent,' is placed on the previously cleaned metal surface. A piece of Perspex sheet (1 in. square × $\frac{1}{8}$ in. thick) is carefully placed on this and pressed down in firm contact with the metal; the excess liquid is extruded round the edges of the Perspex sheet. A mercury vapour lamp, emitting the appropriate ultra-violet radiation, is placed a short distance from the surface and the specimen irradiated for one hour. At the end of this period the liquid is cured and has taken a faithful reproduction of the metal surface. The specimen can be readily removed from the metal. For sloping, vertical, or under surfaces, the drop of monomer–polymer mixture is placed on the Perspex sheet, and this is then pressed on to the desired position and cured. For curved surfaces, such as mill rolls, a strip of Perspex sheet, 1 in. wide, is warmed and bent to the required radius of curvature before the 1-in. squares are cut off.

Talysurf records taken on metal surfaces and their replicas in identically corresponding positions show that faithful reproduction is achieved, even when the records are taken at high magnification. The figures given in Table I also indicate the degree of accuracy which has been attained. Average roughness estimations were carried out on various steel surfaces and on the corresponding replicas. Approximately thirty readings were taken on each specimen.

Table I

AVERAGE ROUGHNESS OF STEEL SPECIMENS AND REPLICAS

The values are given in micro-inches (10^{-6} in.)

Specimen	Steel		Replica	
	Average Roughness	Standard Deviation	Average Roughness	Standard Deviation
1	2·81	0·14	2·81	0·22
2	13·2	1·1	13·5	1·1
3	17·2	0·9	17·2	0·9
4	107·4	7·3	101·4	7·4
5	147	25	146	31
6	195	44	189	41

These results show that a method is now available for assessing surface finish, by means of replicas to be examined on the Talysurf instrument, of any surface, provided that an ultra-violet lamp can be used to irradiate the specimen.

The writers wish to thank Messrs. Imperial Chemical Industries (Plastics), Limited, for their advice and practical assistance and the Coatings Committee of the British Iron and Steel Research Association for permission to publish this letter.

J. PEARSON
M. R. HOPKINS

British Iron and Steel
Research Association,
Sketty Hall Laboratories,
Swansea.

*R. C. Faust and S. Tolansky: *Proceedings of the Physical Society*, 1947, vol. 59, p. 951.

NEWS

ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Symposium on Peeling

A short series of papers dealing with the formation of peel on whiteheart malleable castings is published in this issue of the *Journal*. Arrangements are being made for this subject to form the basis of a symposium, which will take place probably in March. Further details of the meeting will be announced in these columns in the February issue of the *Journal*.

Binding Cases for the Monthly Journal

The binding cases and indexes for the twelve numbers of the *Journal* for 1947, making up volumes 155, 156, and 157, are now in the final stages of preparation, and they will be despatched in the near future to all members who have requested them. A few additional copies of the indexes are available on request and early application should be made to the Secretary.

Bound copies of the three volumes will also be despatched, as soon as they are ready, to all members and others who have already ordered them.

Symposium on Powder Metallurgy

SPECIAL REPORT No. 38

The second (revised) edition of the Symposium on Powder Metallurgy, first published in May 1947, is now being printed. It is expected that copies will be distributed during the next few weeks to those who have already made application.

EDUCATION

National Certificates in Metallurgy

Pass List

The following candidates successfully completed the senior part-time course for the year 1946–47, and have been awarded the Ordinary National Certificate in Metallurgy ; a subject given in parenthesis indicates the award of a distinction in that subject :

Battersea Polytechnic, London—Casling, William Henry ; Graham, John Alistair (General Metallurgy) ; James, Victor Leonard (Chemistry and Physical Metallurgy) ; Proffitt, Henry Joseph.

Birmingham Central Technical College—Burnard, David Chance (Chemistry) ; Lee, Raymond Frederick ; Pester, Dorothy Margaret ; Storer, Roy Alan ; Thomas, John Leslie ; Toach, Shirley Ann ; Warwick, Leslie Vincent.

Chesterfield Technical College—Lynch, John James ; Meakin, John ; Moss, John Barry (Physical Chemistry); Pugh, Jack Fisher ; Sutherland, John McKellar.

Merchant Venturers' Technical College, Bristol—Atherton, Norman James ; Murlis, Douglas Harry ; Phillips, Esme Gladys ; Robson, Alfred Wayman.

Rotherham Technical College—Bond, Leslie ; Broadbent, Peter Robert ; Graves, Geoffrey Basil ; Mayo, George Victor.

Smethwick, The Chance Technical College—Allen, Gordon Leonard (General Metallurgy) ; Brett, Alan Samuel ; Burton, James ; Cornes, Leslie George ; Davis, Derrick Sydney ; Dunning, Malcolm John (General Metallurgy) ; Forsyth, Richard Victor Walter ; Harris, Leslie Howard William (General Metallurgy) ; Harvie, Kenneth Ryder ; Laker, Gordon Patrick ; McManus, William George (General Metallurgy) ; Salmon, Clifford Gerald (Physics) ; Scott, John Edward (Physics and Chemistry) ; Taylor, Alan John ; White, John Harold (Physics and General Metallurgy).

Swansea Technical College—Lewis, Ogwyn; O'Connor, James John ; Rees, George Philip.

Wednesbury County Technical College—Elson, Gladys; Feasey, Raymond James ; Harrison, Jim ; Meadows, Albert William ; Richards, Charles John Godfrey (General Metallurgy) ; Tyler, Arthur William ; Vaughan, Geoffrey Raymond ; Wakeman, Peter John.

Wolverhampton and Staffordshire Technical College—Baker, Joan Theresa ; Crowe, Stanley Frederick ; Firm, Iris ; Fisher, Kevin ; Holgate, Sheila Marjory ; Meacham, Frederick Arthur (General Metallurgy) ; Wicks, Clive William Harry ; Williams, Edward ; Williams, John Randall.

The above candidates have also passed in Engineering Drawing.

The list of successful candidates in the Advanced Part-Time Course for the Higher National Certificate will be published in the February, 1948, issue of the *Journal*.

Metallurgy—A Scientific Career in Industry

A second (revised) edition of the brochure "*Metallurgy—A Scientific Career in Industry*," published in 1946 by the Joint Committee on Metallurgical Education, 4, Grosvenor Gardens, London, S.W.1, is now being prepared.

The first edition of this brochure was found to be of considerable interest both to teachers of metallurgical subjects and to parents with sons interested in metallurgy as a career. The second edition will contain the latest available information with regard to courses in metallurgy at Universities and Technical Colleges.

Seventy-Five Years Ago

Few members will recall events as long ago as seventy-five years and, although the proceedings of The Iron and Steel Institute have been published in the *Journal* from the date of its foundation, only a very limited number of members will be able to refer to the early volumes. It is proposed, therefore, to present in this column a series of extracts from the *Journal* and reviews of the activities of the Institute in 1873. Two meetings of the Institute were held in that year, the Annual General Meeting in April, and a summer meeting held in Liège, in August. For the next six issues of the *Journal* details will be given of the Annual Meeting, whilst in the remaining issues for the year there will be reviews of the papers presented at the summer meeting.

Mr. I. Lowthian Bell was installed as President at the Annual General Meeting held in 1873, succeeding the second President, Mr. Henry Bessemer. The first President, elected four years previously, was the Duke of Devonshire. During the first four years of its life the Institute proved its popularity with the industry, and the speeches at the Annual Meeting showed that every confidence was placed in the ability of the new organization to provide important technical and practical benefits to the iron trades.

The importance of the new Institute was recognized not only in this country, but also in Europe, where the need for a similar organization was becoming urgent, and its high standing and popularity can be judged from the growth of its membership, which was doubled during the first four years. At the beginning of 1873 the register contained 580 names.

In his Presidential address Mr. Lowthian Bell referred to the great benefit to the industry of the new Institute: how its membership was sought, not only by British ironmasters, but by many of their colleagues in Europe and the U.S.A.; and how, by the medium of its meetings and publications, the fruits of research and mutual conference were made available to all, without selfish interest and narrow jealousy.

The President then drew attention to the probable effect of foreign competition which, he said "threatens . . . if not to extinguish, yet seriously to cripple a trade which has occupied a very honourable and conspicuous position in our manufacturing annals. The progress made in other lands . . . has arisen from a rapid adaptation of our appliances, and not from any important discoveries abroad. Until, indeed, we are made acquainted with a series of improvements equal to the puddling process, the substitution of rolling mills for forge hammers, the hot blast, and our last grand achievement of Bessemer steel, we must permit our national pride the luxury of believing that it is to ourselves, more than to any other people, that the world is indebted for the advance which enables it to be furnished with cheap iron."

NEWS OF MEMBERS

▶ Ing. G. BAGNA has changed his business address to Soc. Metallurgica Vittorio Cobianchi, Omegna, Novara.

▶ Monsieur LÉOPOLD BOUVIER has been promoted from Director to Joint General Director at the central offices of Arbed, Luxembourg.

▶ Mr. MALCOLM BROWN has resigned from the Boards of Andrews Toledo, Ltd., Darwins, Ltd., and their associated companies.

▶ Mr. W. F. CARTWRIGHT has resigned from the Board of Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd., on his appointment as Director of the Steel Company of Wales, Ltd. Mr. Cartwright is also General Manager of the Steel Division.

▶ Mr. STANLEY J. DYAL has been appointed a Director of Messrs. Thos. W. Ward, Ltd.

▶ Mr. F. L. ESTEP, consulting engineer, has changed his business address to 25, Broad Street, New York, 4.

▶ Mr. S. H. FREDERICK is now employed as a metallurgist at the Pametrada Research Station, Wallsend-on-Tyne.

▶ Mr. R. H. GARNER, at present Head of the Engineering and Science Department of Blackpool Technical College and School of Art, has been appointed Principal of a new Technical College to be opened in the Burnbank industrial area outside Glasgow.

▶ Mr. N. J. GRIFFITHS has been appointed Works Manager of the Brownhills Works of Cindal Aluminium, Ltd., Lichfield Road, Brownhills, Staffs.

▶ Mr. G. HAIM has relinquished his position with Rockweld, Ltd., and is now associated with Rediffusion, Ltd.

▶ Mr. M. K. HALDAR is shortly returning to India and his address will be: The Indian Association for the Cultivation of Science, 210, Bowbazar Street, Calcutta, 12.

CORRESPONDENCE

Correspondence on Papers published in this issue of the Journal should be submitted to the Editor before 20th March, 1948.

▶ Mr. F. HARDESTY, previously with Messrs. Michie and Davidson, Newcastle, has taken up a position as lecturer in chemistry and metallurgy at Sunderland Technical College.

▶ Mr. E. INESON has resigned his position as chief research metallurgist of Messrs. David Brown and Sons (Huddersfield), Ltd., to take up an appointment in the Metallurgy (General) Division of the British Iron and Steel Research Association.

▶ Mr. P. A. JOB is retiring from his position as General Works Manager of the Llanelly Steel Co., Ltd., but will retain his seat on the Board.

▶ Mr. HAMISH JOHNSTON has changed his address to c/o Briton Ferry Steel Co., Ltd., Briton Ferry, Glam.

▶ Mr. C. R. WHEELER, C.B.E., has been appointed Joint Managing Director, with Mr. THOMAS JOLLY of Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd.

▶ Eng. Rear-Admiral C. W. LAMBERT, C.B., Chief Engineer of the Appleby-Frodingham Steel Co., Ltd., Scunthorpe, has been appointed a Director of that Company.

▶ Mr. D. S. McGREGOR has taken up a post as Senior Assistant in Mechanical Engineering at Darlington Technical College.

▶ Mr. ALEX B. McINTOSH has been appointed in charge of the Metallurgical Department of the Department of Atomic Energy, Risley, near Warrington.

▶ Mr. D. MANTERFIELD has succeeded Mr. B. W. METHLEY, who has retired, as Chief Chemist of Messrs. Steel, Peech, and Tozer.

▶ Mr. T. F. PEARSON has resigned his position with Messrs. Colvilles, Ltd., and is now with the Consett Iron Co., Ltd., Consett, Co. Durham.

▶ Mr. A. D. PHILLIPS has changed his address to c/o Hertford Electric Steel Corporation, 540, Flatbush Avenue, Hertford, Conn., U.S.A.

▶ Mr. E. J. PODE has resigned from the Board of Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd., on his appointment as Managing Director of the Steel Company of Wales, Ltd.

▶ Mr. JOHN RANKINE, formerly at the British Iron and Steel Research Association, Sheffield, is now with the Whitehead Iron and Steel Co., Ltd.

▶ Mr. A. J. NICOL SMITH has returned to his former position of Senior Experimental Officer with the Ministry of Supply (Directorate of Weapons Research (Defence)); his address is Ministry of Supply, W.R.(D.)2, 10th Floor, Shell Mex House, Strand, London, W.C.2.

▶ Mr. W. A. SMITH, of Messrs. Stewarts and Lloyds, Ltd., has now moved from Corby to Clyde Tube Works, Coatbridge, where he is Manager of the Department of Research and Technical Development (Scottish Section).

▶ BALDWINS (Holdings), LTD.—It is much regretted that the statement in the December issue that Col. J. M. BEVAN and Mr. ANDREW AGNEW had become a Director and Vice-Chairman of the Company was incorrect. Apologies are offered to these members for any inconvenience to which this notice may have put them.

Elections

▶ Mr. V. C. FAULKNER has been elected President of the International Committee of Foundry Technical Associations, for 1948.

▶ Mr. FRANCIS B. FOLEY has been nominated for President of the American Society for Metals, the election to be confirmed at the Annual Meeting in October at the National Metal Congress, Chicago.

▶ Mr. ELLIS HUNTER has been re-elected President of the British Iron and Steel Federation.

▶ Mr. V. E. STANBRIDGE has been elected to full membership of the Institution of Structural Engineers.

▶ Dr. CLYDE WILLIAMS has been elected Director of the American Institute of Mining and Metallurgical Engineers, for the year 1948.

Honours and Awards

▶ Mr. W. J. DAWSON, a Vice-President of the Institute, has been awarded the C.B.E. in the New Year's Honours.

▶ Mr. F. FANCUTT, Paint Technologist of the L.M.S. Railway Company, and Dr. J. C. HUDSON, Head of the Corrosion Laboratory of the British Iron and Steel Research Association, have been awarded jointly by the Institution of Civil Engineers a Telford Premium for their paper on "*The Prevention of Corrosion on Structural Steelwork.*"

▶ Mr. T. M. SERVICE, of Wm. Beardmore and Co., Ltd., has been awarded the C.B.E. in the New Year's Honours.

Memoranda

▶ The Wellman Smith Owen Engineering Corporation, Ltd., announce that as from 24th November, 1947, the address of their Head Office in London will be: Parnell House (opposite Victoria Station), London, S.W.1.

Obituary

The Council regret to record the deaths of:

Mr. NORMAN A. CLARK, on 20th November, 1947.

Sanatore GIORGIO ENRICO FALCK, a former Honorary Vice-President, on 12th January, 1947.

Mr. BENJAMIN TALBOT, Past-President, Bessemer Medallist, on 16th December, 1947.

CONTRIBUTORS TO THE JOURNAL

Dr. J. C. Hudson—Official Investigator to the Corrosion Committee and Head of the Corrosion Laboratory of the British Iron and Steel Research Association. Dr. Hudson was born at Plumstead in 1900 and received his early education at Brighton College. Having obtained the B.Sc. degree of London University in 1920, he was awarded a Royal Scholarship at the Royal College of Science; there he studied chemistry and gained a First Class Associateship in that subject in 1923. Following this he carried out two years' post-graduate research under the late Professor J. C. Philip, F.R.S. In 1925 Dr. Hudson was appointed as a second investigator to the Atmospheric Corrosion Committee of the British Non-Ferrous Metals Research Association, and was particularly concerned with the conduct of field tests on non-ferrous metals and alloys under atmospheric conditions. His results, published by the Faraday Society in 1929, formed the basis of a thesis for which he was awarded the Internal D.Sc. degree of London University in the same year.

In 1929 Dr. Hudson accepted an appointment as Official Investigator to the Corrosion Committee of The Iron and Steel Institute and the British Iron and Steel Federation. Since that date he has been responsible for organizing and conducting the work of this Committee on iron and steel corrosion and its prevention. Dr. Hudson has written extensively on the subject of corrosion, and a summary of the investigations will be found in his book "*The Corrosion of Iron and Steel.*"

T. A. Banfield, Ph.D., F.R.I.C.—Research Chemist to the British Anti-Fouling Composition and Paint Co., Ltd. Dr. Banfield was born at Brixham, Devon,

J. C. Hudson

T. A. Banfield

J. S. Bowden

N. H. Bacon

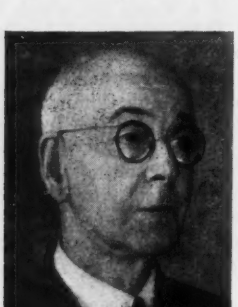
in 1913, and attended the Torquay Grammar School from 1924 to 1930. He continued his studies at the Royal College of Science, South Kensington, until 1934, carrying out research in physical chemistry under the direction of the late Professor J. C. Philip, F.R.S. For a thesis entitled "Chemical Fogs" the University of London awarded him the degree of Doctor of Philosophy in 1934. During the next 12 years he was employed in Birmingham as Assistant Investigator to the Corrosion Committee of The Iron and Steel Institute and the British Iron and Steel Research Association, in which post he gained wide experience in research on the corrosion and protection of iron and steel.

Dr. Banfield joined the British Anti-Fouling Composition and Paint Co., Ltd., in May 1947, where he is in charge of research on ships' bottom compositions.

J. S. Bowden, B.Met., F.I.M.—Senior Assistant in the Development Department of the Tin Research Institute. Mr. Bowden was born in Glossop, Derbyshire, in 1902. He was educated at Glossop Grammar School, where he gained a Derbyshire Major Scholarship, and proceeded to Sheffield University, graduating with the degree of B.Met. in 1923. He took up a business appointment after gaining experience in foundries. In 1942 Mr. Bowden joined the Department of the Superintendent of the Technical Application of Metals, of the Ministry of Supply, as experimental officer in the Midland Region. He transferred to the H.Q. staff in London, in 1944, and acted as officer-in-charge of the S.T.A.M./A.R.D. laboratories at Fort Halstead until February 1947, when he resigned to take up his present position with the Tin Research Institute.

P. Walker—Chief Metallurgist, Cargo Fleet Iron Co., Ltd. Mr. Walker began his career as a junior chemist at the Cargo Fleet Iron Co., Ltd., in 1903. Having had experience in all the firm's laboratories and production departments, he was appointed to his present position in 1926. He joined The Iron and Steel Institute in 1934, and is a member of the Cleveland Institution of Engineers and of the Ingots Committee and the Ingot Surface Defect Sub-Committee of the British Iron and Steel Research Association. He is also representative of the South Durham Steel and Iron Co., Ltd., to the British Iron and Steel Research Association, and a member of the Governing Council of the Constantine Technical College, Middlesbrough.

N. H. Bacon, A.Met., F.I.M.—Chief Works Metallurgist at Messrs. Steel, Peech and Tozer. Mr. Bacon was born in Sheffield in 1892. He joined his present firm in 1911, and was awarded the Associateship in Metallurgy in 1917, after which he served as assistant manager in the basic and acid melting shops for three years, and later in the mills and forges as metallurgical observer. From 1922 to 1928 Mr. Bacon was assistant to the late Mr. E. H. Saniter; he was also in close connection with Mr. J. H. Waslow, from whom he first learned the rudiments of the statistical approach to steelworks problems. Mr. Bacon has written a number of papers on steelworks practice and economic problems.



P. Walker

G. R. Webster

J. Bernstein

J. B R. Brooke

J. S. Bryan

J. C. W. Humfrey

A. E. Probst

G. R. Webster, A.M.I.Mech.E., A.M.I.Met.E., Assoc.I.E.E., F.C.S., F.I.M.—Works Manager of the Britannia Iron and Steel Works, Bedford. Mr. Webster is a native of Birmingham, where he received his early education and served an indentured engineering apprenticeship, later holding both technical and administrative posts in the Midlands and in the London area. He subsequently further widened his experience by spending two years with A.G. der Eisen und Stahlwerke (formerly George Fischer), Switzerland, before becoming associated with his present firm. He is a Member of Council of the Institute of British Foundrymen, and of the Institution of Works Managers, and serves on the Development Committee of the British Cast Iron Research Association.

J. Bernstein, A.R.I.C., A.C.T.C. (Birm.), A.I.M.—Research Metallurgist at the British Cast Iron Research Association, Alvechurch, Birmingham. Mr. Bernstein was educated at King Edward VI Grammar School, Birmingham, and continued his studies at the Birmingham Central Technical College. He gained his Associateship of the Royal Institute of Chemistry in 1944 and was awarded an Associateship of his College, in Chemistry, in the same year. In 1945 he became a Member of the Institute of British Foundrymen, and he was elected an Associate of the Institution of Metallurgists in 1946.

Mr. Bernstein joined the British Cast Iron Research Association in 1937 and was appointed a member of the Metallurgical Research Staff towards the end of 1941.

J. B. R. Brooke, A.M.I.C.E.—Director and General Manager of Messrs. John Lysaght, Ltd., Normanby Park Steel Works, Scunthorpe. Mr. Brooke, who was educated at Charterhouse School, first became associated with the iron and steel trade in 1924 at Frodingham Works. In 1925 he went to Normanby Park Steel Works, serving for some time in most departments. He was appointed Assistant General Manager of these works in 1936, and to his present position in 1943. He spent six months studying the German iron and steel industry in 1928, and six months in America in 1934–35.

Mr. Brooke is a Director of Metallurgical Products, Ltd., and of Ledloy, Ltd., and President of the Lincolnshire Iron and Steel Institute, and of the North Lincolnshire Scientific and Technical Society. He is Vice-Chairman of the Lincolnshire Ironmasters' Association, and a Member of the Steelmaking Divisional Panel of the British Iron and Steel Research Association.

J. S. Bryan — Fuel Engineer at Messrs. John Lysaght, Ltd., Normanby Park Steel Works, Scunthorpe. Mr. Bryan served his apprenticeship with Messrs. John Lysaght, Ltd., from 1933 to 1939. He was elected a corporate member of the Institution of Civil Engineers in 1942. He served as a technical officer in the Royal Air Force from 1943 to 1946. On leaving the R.A.F. Mr. Bryan returned to his present firm to take up a post as technical assistant to the blast-furnace manager. In May 1947 he was appointed assistant development and planning engineer, and to his present position in August 1947.

J. C. W. Humfrey, O.B.E., M.Sc., M.Eng., B.A.—Consulting Engineer with Messrs. Riley, Harbord, and Law. Mr. Humfrey was educated at King's School, Chester, and at University College, Liverpool, and obtained a first-class honours degree in engineering at the Victoria University. After a year's research work under Professor Hele Shaw he was awarded an 1851 Exhibition Scholarship, and worked for two years at Cambridge under Professor J. A. Ewing.

From 1902 until 1909 Mr. Humfrey was metallurgist to Messrs. Baldwins, Ltd., Midland Branch, and was subsequently appointed to the Metallurgy Department of the National Physical Laboratory; as chief analyst at the Admiralty Inspection Offices, Sheffield, from 1913 to 1919, when he was awarded the O.B.E.; as metallurgist to Messrs. Sandberg, Consulting and Inspecting Engineers; as Assistant Superintendent, later becoming Deputy and finally Superintendent of the Technical Application of Metals, Ministry of Supply. He joined the firm of Messrs. Riley, Harbord, and Law in 1947.

A. E. Probst—Foundry Manager at the Britannia Iron and Steel Works, Ltd., Bedford. Mr. Probst served his apprenticeship in the foundry departments of the Eisenwerk, Klus, Switzerland, of the Von Roll concern, and having obtained a diploma in mechanical engineering at the Technikum Burgdorf, Switzerland, he then spent a short period in the drawing office of Theodor Bell and Co., Kriens, Lucerne. On leaving Lucerne he joined his present firm as assistant foundry manager.

IRON AND STEEL ENGINEERS GROUP

Forthcoming Meetings

London

The **Sixth Meeting** of the Iron and Steel Engineers Group will be held at 4, Grosvenor Gardens, London, S.W.1, on Wednesday, 18th February, 1948.

At the morning session (10.30 A.M. to 12.45 P.M.)

a paper on "*Gaseous and Liquid Fuels in Iron and Steel Works*," by J. B. R. Brooke and J. S. Bryan (J. Lysaght, Ltd.) will be presented and discussed.

At the afternoon session (2.0 P.M. to 4.0 P.M.) Mr. H. Southern, Chairman of the Society of Furnace Builders, will present a paper by that Society on "*Soaking Pits*," which will also be discussed.

The two papers will be found in this issue of the *Journal*.

Provinces

Further meetings of the Iron and Steel Engineers Group have been arranged as follows:

30th January—Joint meeting with the West of Scotland Iron and Steel Institute, at the Society of Engineers and Shipbuilders, 39, Elmbank Crescent, Glasgow, at 6.45 P.M. Mr. J. B. R. Brooke (J. Lysaght, Ltd.) will present a paper on "*Gaseous and Liquid Fuels in Iron and Steel Works*" (published in this issue of the *Journal*).

21st February—Joint meeting with the Swansea and District Metallurgical Society, at the Royal Institution of South Wales, Swansea, at 6.30 P.M. Mr. W. F. Cartwright (Steel Company of Wales, Ltd.) will read a paper on "*Engineering Aspects of a Modern Steelplant*."

4th March—Joint meeting with the Leeds Metallurgical Society, at the Chemistry Department, Leeds University, at 7 P.M. Dr. A. T. Bowden (C. A. Parsons and Co., Ltd.) will present a paper on "*Gas Turbine Applications in Iron and Steel Works*" (published in the May, 1947, issue of the *Journal*).

THE BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Oil Metering

A colloquium was held at The Iron and Steel Institute on Thursday, 30th October, 1947, when the Plant Engineering Instruments Sub-Committee met, together with representatives of oil companies, meter manufacturers, and Imperial Chemical Industries, Ltd., to discuss the problems set out in the report by Dr. Bramley on "Metering of Liquid Fuel Flow—(*a*) Heavy Fuel Oils." The possibility of using a metering pump was a point of particular interest during the discussions. Several available types of pumps have been considered and it is hoped soon to instal at least one type to be used for extended tests in a melting shop as well as to carry out laboratory tests.

Junior Steelmaking Conference

The First Junior Steelmaking Conference was held at Ashorne Hill on Wednesday and Thursday, 26th and 27th November, 1947. Organized for the purpose of providing opportunities for discussion and conference to those members of the staffs of member firms who are not able to attend the normal conferences, this first experimental meeting was a complete success.

Under the Chairmanship of Mr. A. Robinson of the Appleby-Frodingham Steel Company the conference opened with the discussion of a paper on "Steel Defects and Their Causes" by Mr. N. H. Bacon of Steel, Peech & Tozer. This was followed by a short discussion on general steelmaking topics introduced by the Chairman, who drew attention to some of the urgent problems now facing the industry.

The first session on the second day was devoted to the discussion of a paper on "The Control of Open-Hearth Furnaces by Instruments," which was presented by Dr. J. H. Chesters of the United Steel Companies, Ltd.

The lively discussion provoked by these papers and the obvious enthusiasm of the speakers indicated that further such conferences would be most successful.

AFFILIATED LOCAL SOCIETIES

Ebbw Vale Metallurgical Society

Affiliation

The Council of the Ebbw Vale Metallurgical Society has accepted the Institute's invitation to become affiliated. This arrangement takes effect as from 1st January, 1948.

Joint Meeting

At the kind invitation of the Ebbw Vale Metallurgical Society a Joint Meeting of that Society with The Iron and Steel Institute will be held in the Rational Hall, Ebbw Vale, on Tuesday, 9th March, 1948, at 6.45 P.M. Mr. E. Hemingway-Jones, Chairman of the Ebbw Vale Metallurgical Society, will preside, and Dr. A. Parker, C.B.E. (Director of Fuel Research, Department of Scientific and Industrial Research), will give a lecture on "*Coal and the Future*."

Details of other meetings of the Affiliated Local Societies, to which members of the Institute are invited, will be found in the Diary.

Special Report No. 39

By arrangement between the Council of The Iron and Steel Institute and the Councils and Committees of the Affiliated Local Societies, selected papers presented to these societies are being published by the Institute in the Special Report series. The first of these reports, Special Report No. 39, is now being printed and copies will be sent to the Secretaries of the Societies for distribution to their members.

Members of the Institute may obtain copies at the reduced rate of 5*s*. each on application to the Secretary. Non-members of either the Institute or any of the Societies may obtain copies at the cost of 10*s*.

INSTITUTION OF METALLURGISTS

Registrar-Secretary

The Institution of Metallurgists requires a full-time Registrar-Secretary and applications for the appointment are invited. The post is suitable for men with first-class qualifications in metallurgy or other allied sciences, who have had administrative experience and who are keenly interested in the advancement of the profession. The initial salary will be not less than £1000 p.a. Applications with full particulars, which will be treated with the strictest confidence, should be addressed to the President at 4, Grosvenor Gardens, London, S.W.1.

NEWS OF SCIENCE AND INDUSTRY

British Welding Research Association

A meeting of the Nominees of Member Firms, Members of the Council, and the Welding Research Board, of the British Welding Research Association, was held at the Offices of the Association, 29, Park Crescent, London, W.1, on Tuesday, 18th November, 1947, at 3 P.M.

After a welcome extended to the guests by Sir Stanley V. Goodall, K.C.B., O.B.E., Chairman of Council, Sir William J. Larke, K.B.E., President of B.W.R.A., addressed the meeting, and in conclusion said that the Association, by the speedy solution of its investigations and the increased industrial application of its results, could thereby increase the competitive powers of British industry.

Dr. H. G. Taylor, Director of Research of the Association, in the course of his remarks, outlined the present work of the Association and its future policy, and emphasized the importance which he attached to personal contact with the Member Firms.

Sheet and Strip Rolling

A course of six lectures will be given in the School of Industrial Metallurgy, University of Birmingham, on "*An Introduction to the Theory of Sheet and Strip Rolling, and Its Application to Industrial Problems*," by Dr. A. R. E. Singer. These lectures, which will take place on successive Saturday mornings, commencing on 31st January, 1948, at 10.30 A.M., are planned to meet the requirements of senior executive officers, metallurgists, engineers, and technicians in the metal-producing and metal-consuming industries. The syllabus will cover theories applicable to the rolling of strip and sheet, and the factors governing the development and choice of rolling-mill plant.

Quantum Chemistry

A course of six lectures on "*Quantum Chemistry*," by Professor H. S. W. Massey, F.R.S., will be held on Tuesday evenings, from 6 to 7 P.M., beginning on Tuesday, 3rd February, 1948, at the Sir John Cass Technical Institute, Jewry Street, Aldgate, London, E.C.3. The course is intended for post-graduate and advanced students and will be addressed chiefly to those who do not possess the mathematical equipment which is necessary for the usual treatment of these problems. The syllabus of the course includes the electronic structure of atoms, quantum theory, crystal binding, nuclear structure, etc. Applications for enrolment should be made to the Principal.

British Ceramic Research Association

As from the beginning of this year, the two hitherto separate bodies associated with ceramic research—the British Refractories Research Association, and the British Pottery Research Association—will be united under the title of the British Ceramic Research Association.

CORRIGENDA

Discussion on Special Report No. 36

The following correction should be noted to the "*Discussion on a Symposium on the Hardenability of Steel*," which was published in the September, 1947, issue of the *Journal*: Page 67, left-hand column, line 18, "Fig. 42" should read "Fig. 22."

Contributors to the Journal—The late Dr. Jay

In the biographical notes on the late Dr. A. H. Jay, on page 282 of the October, 1947, issue of the *Journal*, the date when Dr. Jay joined the United Steel Companies, Ltd., should be October, 1933, instead of October, 1943.

Correspondence on Contamination of Thermocouples

The written contribution by Mr. W. J. Todd, printed on pages 197 and 198 of the October, 1947, issue of the *Journal*, refers to the paper "*A Symposium on the Contamination of Platinum Thermocouples*," by the Liquid Steel Temperature Sub-Committee, and not to the paper by Mr. J. Hall, which was discussed jointly with the paper by the Liquid Steel Temperature Sub-Committee.

It should also be noted that the sentence on lines 4 and 5 of the second paragraph of Mr. Todd's contribution should read "The present series of investigations are concerned practically solely with breakage due to brittleness, and a completely satisfactory explanation has been established" and not "The present series of investigations afford a completely satisfactory explanation."

Scaling Behaviour of Heat-Resisting Steels

In the paper by W. Stauffer and H. Kleiber, on "*The Scaling Behaviour of High-Strength Heat-Resisting Steels in Air and Combustion Gases*," published in the June, 1947, issue of the *Journal*, the ordinates in Figs. 11 to 20, pages 186–188, should read "Scaling loss, g./sq. m./hr." and not "Scaling loss, g./sq. in./hr."

Special Report No. 30

In Special Report No. 30, "*Ironmaking at the Appleby-Frodingham Works of the United Steel Companies, Ltd.*," by G. D. Elliott and others, the following correction is made to certain calculations in Section VI: Owing to a mathematical error of multiplying by 2 in calculating the modulus of rigidity of blast-furnace stove refractories, the figures (in lb./sq. in. $\times 10^5$) quoted in Table LXVIII and Fig. 75 are double their correct value. The authors state, however, that the absolute value of the rigidity modulus is only of secondary importance to the changes occurring on firing and in service in the stove, and that their theories and conclusions in this Section of the Report are in no way affected.

Modified Type of 405 Stainless Iron

In the text of the abstract of the above paper by M. A. Scheil, on page 310 of the October, 1947, issue of the *Journal*, the modified type of stainless steel is given as "450" instead of "405."

Abnormal Creep in Carbon Steels

The following correction to Mr. W. B. Brook's contribution to the discussion on the paper "*Abnormal Creep in Carbon Steels*," by J. Glen, which was published in the December, 1947, issue of the *Journal*, should be noted: On page 579, second paragraph, line 14, "0·00006 in./in./hr." should read "0·00006 %/hr."

DIARY

22nd January

MIDLAND METALLURGICAL SOCIETIES — Symposium on Chain-Making—James Watt Institute, Great Charles Street, Birmingham, 6.30 P.M.

23rd January

SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—Joint Meeting with the Institute of Metals (Sheffield Section), "*Some Metallurgical Problems in the Field of Atomic Energy*," by E. W. Colbeck—Royal Victoria Station Hotel, Sheffield, 6.15 P.M.

27th January

LINCOLNSHIRE IRON AND STEEL INSTITUTE—"*Oil-Firing of Open-Hearth Furnaces*," by S. W. Evans—Technical School, Cole Street, Scunthorpe, 7.30 P.M.

30th January

WEST OF SCOTLAND IRON AND STEEL INSTITUTE—Joint Meeting with the Iron and Steel Engineers Group, The Iron and Steel Institute, "*Gaseous and Liquid Fuels in Iron and Steel Works*," by J. B. R. Brooke—Society of Engineers and Shipbuilders, 39, Elmbank Crescent, Glasgow.

2nd February

CLEVELAND INSTITUTION OF ENGINEERS—"*Some Aspects of Modern Blast-Furnace Operation*," by R. P. Towndrow—Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, 6.30 P.M.—Chairman, C. Moody.

5th February

LEEDS METALLURGICAL SOCIETY—"*The Investigation of Failures*," by C. P. Miller—Chemistry Department, Leeds University, 7.0 P.M.—Chairman, A. Preece.

5th February

MIDLAND METALLURGICAL SOCIETIES — Joint Meeting with The Iron and Steel Institute, "*Modern Developments in Hardenability Testing*," by W. Steven—James Watt Institute, Great Charles Street, Birmingham, 6.30 P.M.

6th February

MANCHESTER ASSOCIATION OF ENGINEERS—"*Fuel Efficiency*," by A. C. Dunningham—Engineers' Club, Albert Square, Manchester, 6.45 P.M.

10th February

EBBW VALE METALLURGICAL SOCIETY—"*Stainless Steel*," by C. R. Pipe—Lesser Workmen's Hall, Ebbw Vale, 7 P.M.

16th February

SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—Exhibition of Technical Films—Royal Victoria Station Hotel, Sheffield, 6.15 P.M.

18th February

IRON AND STEEL ENGINEERS GROUP, THE IRON AND STEEL INSTITUTE—Sixth Meeting—4, Grosvenor Gardens, London, S.W.1, 10.30 A.M.

MIDLAND METALLURGICAL SOCIETIES

The programme for the meeting of the above Societies to be held on 22nd January, was incorrectly advertised in last month's *Journal*; the correct version is given in this issue.

TRANSLATION SERVICE

(The previous announcement was made in the December, 1947, issue of the *Journal*, page 628).

TRANSLATIONS AVAILABLE

No. 329 (German). A. GELEJI: "Calculating the Forces Arising and the Power Requirements in the Mannesmann Tube-Rolling Process." (*Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy*, 1941, vol. 13, pp. 208–223).

No. 330 (German). A. GELEJI: "The Theoretical Problems Involved in the Design of Rolling-Mill Stands." (*Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy*, 1941, vol. 13, pp. 224–242).

PAPERS ACCEPTED FOR TRANSLATION SINCE THE LAST ANNOUNCEMENT

(German). H. BABLIK and F. GÖTZL: "The Lead-Zinc Galvanizing Process." (*Metallwirtschaft*, 1942, vol. 21, Nov. 27, pp. 718–719). H. BABLIK: "The Formation of Hard Zinc in the Lead-Zinc Galvanizing Process." (*Stahl und Eisen*, 1944, vol. 64, Feb. 24, pp. 120–122).

(Swedish). W. KARLSSON: "The Possibilities of Using Liquid Fuels in Iron and Steel Works." (*Värmlandska Bergsmannaföreningens Annaler*, 1946, pp. 47–94).

(German). F. LÜTH: "The Moisture in Industrial Gases. Part II—Application of the Moisture Calculation." (*Archiv für das Eisenhüttenwesen*, 1930, vol. 4, Oct., pp. 185–192).

(Dutch). M. G. VAN DER STEEG: "The Heat-Treatment of Chains." (*Metalen*, 1947, vol. 1, June, pp. 190–192; July, pp. 207–209).

CHARGES FOR COPIES OF TRANSLATIONS.—For the above translations a charge will be made of 10*s*. for the first copy and 5*s*. each for each additional copy of the same translation. Requests should be accompanied by a remittance. The above translations are not available on loan from the Joint Library.

TRANSLATIONS PREPARED AT MEMBERS' REQUEST.—Members requiring translations of foreign papers are invited to communicate with the Secretary, who will ascertain whether the translations can be prepared for inclusion in the series.

ABSTRACTS OF CURRENT LITERATURE *and* BOOK NOTICES

IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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FUEL—PREPARATION, PROPERTIES, AND USES

The Occurrence of Sulphur and Nitrogen in Coal. L. Horton and R. B. Randall. (XIth International Congress of Pure and Applied Chemistry: Fuel in Science and Practice, 1947, vol. 26, Sept.-Oct., pp. 127–132). A brief review is made of present knowledge of the organic sulphur and nitrogen compounds in coal and an account is given of an investigation in progress on the distribution of nitrogen and sulphur in a bright coal.—C. O.

The Determination of Moisture in Solid Fuels. B. G. Simek and J. Ludmila. (Fuel in Science and Practice, 1947, vol. 24, Sept.-Oct., pp. 132–137). The results of determinations of the water content of solid fuels by distillation methods are compared with those obtained by direct absorption in calcium chloride after drying in a stream of nitrogen. The authors conclude that the best methods for the determination of water are (*a*) xylene-distillation and (*b*) direct determination by drying at 130° C. in a stream of nitrogen.—C. O.

X-Ray Diffraction Measurements on Solid Fuels. M. Patry. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 263–269). From the study of a large number of solid fuels, such as wood, peat, lignites, coals, cokes, and graphite, the author classifies solid fuels in three classes: graphitic, graphite-cellulosic, and cellulosic. The graphitic fuels, which comprise all coals and artificial fuels, as well as some very old lignites are the most important.—R. F. F.

New Dry-Cleaning Plants at Holmside. C. W. H. Holmes. (Transactions of the Institution of Mining Engineers, 1947, vol. 106, July, pp. 574–592: Colliery Guardian, 1947, vol. 175, July 25, pp. 105–107; Aug. 1, pp. 139–142). An illustrated account is given of the design and performance of the coal-cleaning plants installed by the Holmside and South Moor Collieries, Ltd., Co. Durham, at the Craghead Busty and Oswald pits.—R. E.

Progress in Coal-Preparation Practice. A. Grounds and W. Reed. (Transactions of the Institution of Mining Engineers, 1947, vol. 106, May, pp. 413–452). A survey is made of the coal-cleaning plants in the United Kingdom. The number of plants installed annually since 1890, the number of plants built in the South Wales, Yorkshire, Durham, and North Midlands coalfields from 1920 to 1946, and the number of plants erected in the various districts, classified according to age, together with the rated capacity of each, are shown. Dense-medium processes are making considerable headway, and it is estimated that 9,880,000 tons of coal are cleaned annually by the dense-medium plants in operation. Attention has been paid during recent years to the development of a shallow dense-medium bath in place of the deep type common to Barvoys, Chance and Tromp plants. The advantages and disadvantages of the shallow bath are pointed out, and the Ridley-Scholes plant is described. The trend of coal-cleaning practice on the Continent is considered, and de-watering coal, dust-proofing, and the relative advantages of central and individual plants are discussed.—R. E.

A Review of Coal Utilization in Australia. I. McC. Stewart. (Journal of the Institution of Engineers, Australia, 1947, vol. 19, May, pp. 97–112).

Liaison in the Iron and Steel Industry, between the Producing and Consuming Centres of Thermal Energy. C. Berthelot. (Chimie et Industrie, 1941, vol. 46, July, pp. 17–20). In order to save fuel, the maintenance of close liaison between producers and consumers of

energy, usually through the use of a fuel-control centre, is suggested.—R. F. F.

Free Expansion Ovens. G. Godin. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 11, pp. 554–556). The new coke-oven built on the principle of free-expansion, by the Houillères Nationales de Pont-à-Vendin, is described, and results obtained are discussed.—R. F. F.

Modernization of French Coke-Ovens. D. Petit. (Revue de l'Industrie Minérale, Mémoires, 1947, Oct., pp. 532–539). The construction of coke-ovens is discussed from the point of view of French reconstruction, with special reference to the possibilities of ovens with free expansion.—R. F. F.

New Coke-Oven Plant for South Africa. (Iron and Coal Trades Review, 1947, vol. 155, Nov. 14, pp. 936–938). Details are given of a complete new coke-oven plant to be built on a virgin site at Vanderbijl Park, South Africa, as part of the expansion programme of the South African Iron and Steel Industrial Corporation, Ltd. The installation will consist of a battery of 54 Woodall-Duckham Becker coke-ovens, together with by-product plant and ancillary equipment; it is rated for the carbonization of 1151 tons of coal per day. An outline is also given of the Corporation's existing plant at Pretoria.—J. R.

The Economic Site for a Coke-Oven Plant. J. Sárek. (Hutnické Listy, 1947, vol. 1, No. 11, pp. 241–243). [In Czech]. The advantages of placing the coke-oven plant close to the blast-furnace are discussed.—R. A. R.

Foundry Coke Quality Effect on Cupola Melting. D. E. Krause. (American Foundryman, 1947, vol. 12, Sept., pp. 44–49). The effect of changes in the quality of coke on cupola operation and methods of getting the best out of the coke available are discussed.—R. A. R.

Centralized Gas Generation—Some Experiences at the Degerfors Steelworks. E. Hermanson. (Jernkontorets Annaler, 1947, vol. 131, No. 10, pp. 425–464). [In Swedish]. Mechanically operated gas producers of the low-temperature type consist of two sections, the coal being carbonized at low-temperature in the upper section by a portion of the producer-gas made from the coke in the lower water-jacketed section. The producer has two gas off-takes, one at each section. The maximum diameter is 8 ft. 6 in., and the producer is able to gasify up to 3500 lb. of coal/hr., equivalent to 60 lb./sq. ft./hr. There are three such producers at the Degerfors steelworks, one used for supplying hot crude gas, from which dust has been removed, to a 50-ton basic open-hearth steel furnace. The other two producers are each equipped with an electrostatic tar separator. They supply cool clean gas to reheating furnaces in the rolling mills. The fuel now used at Degerfors consists of 40–85% Polish or American coal, generally of a rather unsuitable quality for most producers, the rest being wood chips or a mixture of wood chips, peat, and coke-breeze. The calorific values of the gases produced from the normal coal-wood mixture are: (*a*) Low-temperature gas (from upper off-take) 180–190 B.Th.U./cu. ft.; (*b*) hot gas (from lower off-take) 150–160 B.Th.U./cu. ft.; and (*c*) mixed gas about 170 B.Th.U./cu. ft. The following advantages have been noted as compared with ordinary producers: (1) Inferior fuels unsuitable for other types of producers have been used with good economy; (2) the producer has a very high rating and a good yield; (3) the heat value of the gas is higher; and (4) saving in labour. A disadvantage is the difficulty in neutralizing phenols obtained from the tar separators.—R. A. R.

The Thermal Efficiency of Gas Producers. A. C. Dunningham. (Institute of Fuel, 1947, Preprint). The results of efficiency trials for a variety of gas producers at different works are presented. A standard definition of the thermal efficiency of a producer is suggested, and it is shown that the hot-gas efficiences of all producers should be not less than 85%. Methods of increasing cold-gas efficiency are indicated. The author shows that the cost of a therm in hot producer gas is more than 1·5 times that of a therm from mechanically stoked bituminous coal and he therefore recommends that solid fuel should be used whenever possible.—C. O.

A Note on Variations in Producer-Gas Quality. J. E. de Graaf. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 183–190). An account is given of some factors—including the tar content and humidity—and their variations which influence the quality of producer gas. The quality of the gas is improved by high volatile-matter content, high fusion point of the ash, etc., but especially by avoiding fines in the coal. The operation of the producer is also considered, and the influences of throughput, channelling, etc., are ascertained. Approximate rules are given for adapting the steam/air ratio to varying conditions, *e.g.*, properties of the coal.

REFRACTORY MATERIALS

The Mineralogical Composition of Magnesite Bricks. G. R. Rigby, H. M. Richardson, and F. Ball. (Bulletin of the British Refractories Research Association, 1946, No. 71, June: Transactions of the British Ceramic Society, 1947, vol. 46, Oct., pp. 313–329). The results of X-ray analyses of a number of specimens belonging to the five-component system MgO–Fe_2O_3–Al_2O_3–CaO–SiO_2 are presented. A number of rules have been formulated from these results by which it is possible to estimate the mineralogical composition of magnesite products from a knowledge of their chemical composition—C. O.

Investigations of Raw Materials for the Refractory-Quartzite Industries. W. Ahrens and J. H. Hellmers. (Technik, 1947, vol. 2, Sept., pp. 420–422). It is pointed out that it is the structure of quartzite which determines whether it can be used in the refractory industry. The occurrence and types of quartzite are discussed, and a new method of investigating its structure and composition and quantitatively determining its suitability for refractory purposes is described.—R. A. R.

The Raw-Materials Basis of the Refractory Products Industry. J. Kratzert. (Stahl und Eisen, 1947, vol. 66–67, Sept. 11, pp. 321–325). The availability of raw materials for making steelworks refractories in the zones of occupation in Germany are reviewed. The position in the British zone is particularly unfavourable.—R. A. R.

Elastic Constants of Magnetite, Pyrite, and Chromite. M. S. Doraiswami. (Proceedings of the Indian Academy of Sciences, 1947, vol. 25, Section A, May, pp. 413–416). A short account is given of the determination of the elastic constants of magnetite, pyrite, and chromite by the method of Bhagavantam and Bhimasenachar.—C. O.

Basic Ports and Roofs in European Openhearth Furnaces. P. Bettembourg. (Metal Progress, 1947, vol. 52, Sept., pp. 403–410). An account is given of experience on the Continent, and particularly in France, with the use of "Radex E" chrome-magnesite bricks in the construction of open-hearth ports and roofs. Lives of up to 1200 heats have been obtained with self-supporting insulated roofs of 15-ft. span.—C. O.

The Spalling Resistance of Magnesite Products. Part 1—Preliminary Trials on Grading a Sea-Water Magnesite. W. F. Ford and W. J. Rees. (Bulletin of the British Refractories Research Association, 1944, No. 67, June: Transactions of the British Ceramic Society, 1947, vol. 46, Sept., pp. 290–297). An account

is given of preliminary tests on the effect of grading on the spalling of magnesite bricks. The highest resistance to thermal shock of a dead-burned sea-water magnesite was given by a mixture containing equal parts of coarse and fine material, and no material of intermediate grain-size. In general, spalling resistance decreased considerably with more than 10% of the intermediate grade present. Increased firing temperature and the addition of up to 7·5% of iron oxide in the mix may reduce susceptibility to thermal shock.—C. O.

The Spalling Resistance of Magnesite Products. Part II—Experiments on the Effect of the Grading and Grain Shape of the Batch and of the Temperature of Firing. W. F. Ford and W. J. Rees. (Bulletin of the British Refractories Research Association, 1945, No. 69, June : Transactions of the British Ceramic Society, 1947, vol. 46, Sept., pp. 297–312). A detailed account is given of tests of the effect of grading, grain shape, and firing temperature on the spalling resistance of bricks made from two types of sea-water magnesite, a Manchurian magnesite, and an Austrian magnesite. The results indicate : (1) Minimum porosity and maximum spalling resistance are obtained from batches containing equal quantities of coarse and fine grades ; (2) there is a straight-line relationship between percentage apparent porosity and the coefficient of angularity of the grains ; (3) the greater the angularity of the material the less is its spalling resistance, because of the higher porosity and a tendency to friability.—C. O.

Basic Bottoms of Open-Hearth Furnaces and Electric Furnaces. M. Guédras. (Métallurgie, 1947, vol. 79, May, pp. 11–12). After a brief discussion of basic refractory materials used in France, the author gives a short account of the construction of basic open-hearth furnace bottoms.—J. C. R.

The Problem of Refractories in the Open-Hearth Furnace. G. Grenier. (Revue de l'Industrie Minérale, 1944, Comptes Rendus, Feb., pp. 9–13 ; Apr., pp. 23–28). Properties, testing, and uses of refractories for open-hearth furnaces are reviewed.—R. F. F.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Research in Iron Making. T. Bishop. (Iron and Coal Trades Review, 1947, vol. 155, Oct. 31, pp. 835–845). A comprehensive account is given of the continuous co-operative research work being carried out on iron-making problems by the Ironmaking Division of the British Iron and Steel Research Association, the British Cast Iron Research Association, the British Refractories Research Association, and various industrial organizations and academic bodies. Among the items discussed are (1) preparation of the burden, (2) production of sinter, (3) distribution in the blast-furnace, (4) gaseous and non-metallic inclusions in pig-iron, (5) chemical reactions, (6) blast-furnace design and practice, (7) refractories, and (8) investigation of blown-out furnaces. Details are also given of the $\frac{1}{8}$-th scale model blast-furnace at the Stoke works of the Shelton Iron, Steel and Coal Co., Ltd., which, although operated along production lines, is essentially a piece of experimental apparatus. —J. R.

Outline of the Economical Conditions in Metallurgical Production. J. Sárek. (Hutnické Listy, 1947, vol. 2, No. 2, pp. 25–30). [In Czech]. The siting and layout of old and modern ironworks are considered. The economic factors which influence the selection of the blast-furnace site, whether adjacent to the ore or the coal, are given special attention, and the conclusion is that it is advantageous to build the furnaces on the coalfield.—R. A. R.

The Surigao Iron Ore and Its Metallurgical Treatment. E. Ostrea. (Philippine Geologist, 1946, vol. 1, Dec., pp. 5–19). A very short account is given of the laterite ore deposit in Surigao Province, Mindanao, which is said to contain 500 million metric tons of ore, with an average iron content of 47%. The author suggests the method by which the ore should be smelted in the blast-furnace, using the similar Mayari ore in Cuba as an analogy.—C. O.

New Techniques in the Iron and Steel Industry. Blast-Furnaces and Open-Hearth Furnaces. G. Grenier. (Echo des Mines, 1947, Aug., pp. 130–131 ; Sept., pp. 145–146). New practices for crushing ores, the time of stock descent, carbon monoxide concentration and blast pressure in the blast-furnace, and oxygen-enrichment of combustion air or the introduction of pure oxygen in the open-hearth furnaces, are discussed.—R. F. F.

What the Diagram Tells. J. Sárek. (Hutnické Listy, 1947, vol. 2, No. 1, pp. 4–7). [In Czech]. The author evaluates the difference in the consumption of coke and in manpower when producing pig iron from rich and lean ores.—R. A. R.

Jacking Up a Blast Furnace Shell. J. Neary. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. 79–82). A detailed account is given of the procedure followed at Sparrows Point, Maryland, for straightening-up a blast-furnace which was 5 in. out of plumb at the mantle. The operation was carried out while the shell was being renewed.—C. O.

Chemical Removal of Scale, Sludges and Oxides from Steel Plant Equipment. B. H. McDaniel. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. 43–49). The use of chemical solvents for the removal of scale and sludge deposits from water-cooling and similar equipment in heavy industry has considerably increased in the past five years owing to the development of inhibitors which are effective at elevated temperatures. Details are given of the operations involved in removing iron oxides, silicates, clay, organic matter, copper salts, and calcium carbonate from the water-cooling system of a blast-furnace, while in service.—C. O.

Reducing Maintenance of Collector Assemblies. C. E. Baxter. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. 35–37, 85). Improvements in the design of mechanical equipment which have been made to reduce the amount of maintenance required in a blast-furnace and coke-oven plant are described. The examples illustrated are a redesigned current collector for electrically driven ore-transfer cars, the replacement of the porcelain insulators on an outdoor power rail by paraffin-impregnated maple-wood blocks, and the replacement of a motor used for closing contacts on an ore-pushing equipment by a solenoid.—C. O.

Some Relations in the Composition of Blast-Furnace Slags. A. N. Ramm. (Stal, 1939, vol. 9, No. 3, pp. 13–17 ; Chem. Abs., 1939, vol. 33, p. 9227). The composition of 37 blast-furnace slags from the U.S.A., Germany, England, and Russia corresponded very closely to the equation $(RO) = 50 - 0{\cdot}25\,(Al_2O_3) + 3\,(S) - [(0{\cdot}3\,[Si] + 30\,[S])/A]$ where (RO) is the combined percentage of CaO, MgO, MnO, and FeO in the slag, (Al_2O_3) and (S) are the percentages of Al_2O_3 and sulphur in the slag, [Si] and [S] are the percentages of silicon and sulphur in the metal, and A is the ratio of the weight of the slag to the weight of the metal. The equation held for furnaces producing 12%-silicon pig iron and spiegeleisen as well as for open-hearth, Bessemer, and foundry iron, and can be applied more generally than other equations for calculating the charge.—R. A. H.

Relation between Slag Temperature and Composition of Pig Iron in the Blast-Furnace Process. N. U. Yakubtsiner. (Stal, 1940, vol. 10, No. 1, pp. 1–9 ; Chem. Abs.,

1941, vol. 35, p. 3202). In the manufacture of pig iron containing 0·35–1·2% of silicon and 0·04–0·1% of sulphur the temperature of the slag tapped before the pig iron varied between 1320° and 1500° C. Under normal conditions with a slag having a $CaO : SiO_2$ ratio of 1 : 1, an increase in slag temperature of 18° C. will increase the silicon by 0·1%, and an increase of 26° C. will decrease the sulphur by 0·01%.—R. A. H.

PRODUCTION OF STEEL

War Time Expansion of Carnegie-Illinois Steel Corporation in the Pittsburgh District. T. J. Ess. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. C-I-13–C-I-32). A detailed illustrated description is given of the new iron and steel plant built by the Carnegie-Illinois Steel Corporation near Pittsburgh during the war. The main expansion was at Homestead, where a new open-hearth shop, containing eleven 225-ton basic furnaces, a slabbing mill, a plate mill, and a forge shop were built. At the Duquesne works, three basic electric arc furnaces and new heat-treating facilities were installed, and at the Edgar Thomson works two furnaces with hearths 27½ ft. in dia. were constructed.—C. O.

Steel Research Helps Meet Needs of Industry. C. M. Parker. (Materials and Methods, 1947, vol. 26, Aug., pp. 65–70). A survey is made of some of the results of investigations made by research teams in the United States on subjects related to the production of iron and steel.—C. O.

Die Steels and Their Treatments Pace Drop Forge Industry. A. F. Finkl. (Metal Progress, 1947, vol. 52, Oct., pp. 581–584). The development of die steels by the firm of Messrs. A. Finkl and Sons, Ltd., Chicago, over the last thirty years is reported. The trend has been away from plain carbon and nickel and chromium steels towards the triple-alloy, nickel-chromium-molybdenum, steels. Great advances have also been made in heat-treatment techniques.—C. O.

Blowing High Manganese Metal in a Side Blown Converter. J. H. Hall. (Foundry, 1947, vol. 75, Oct., pp. 85, 232). It has been considered impossible to blow metal containing 1% or more of manganese in the side-blown converter owing to the very violent boiling which occurs. The author's theory was that if a slag could be produced early in the blow containing MnO 60% and SiO_2 40%, it would be continuously fluid, and would react with the carbon of the bath at frequent intervals without any violent boiling. The economic advantage of such a process was that foundries could use up their own high-manganese scrap. Details are given of cupola charges, logs of blowing, and analyses of the steel produced, which proved that the theory could be applied with success.—R. A. R.

Venturi Furnaces. R. B. Lane. (Iron and Coal Trades Review, 1947, vol. 155, Oct. 10, pp. 701–703, 704). Recent trends and possible future developments in the design and operation of open-hearth furnaces are discussed, and the advantages of venturi-type furnaces, especially in connection with the adoption of oil-firing and the introduction of oxygen into the bath and incoming fuel, are outlined.—J. R.

Design of All-Basic Open Hearth. (Iron and Steel Engineer, 1947, vol. 24, Oct., pp. 78–83 : Steel, 1947, vol. 121, Nov. 3, pp. 112–118). A detailed illustrated description is given of the construction of an experimental 122-ton open-hearth steel furnace at the South Chicago works of the Carnegie-Illinois Steel Corporation. All parts of the furnace exposed to high temperatures, except the checker chambers and slag pocket, are of basic refractories—mostly metal-encased or plain chemically bonded chrome-magnesite, or burned chrome-magnesite, of varying magnesite content. Both the main and the port roof are of the suspended-arch type, and the main roof is some 8-in. higher than the silica one which has been replaced.—C. O.

Installation and Use of Instruments on Open-Hearth Melting Furnaces. R. C. Baker. (Journal of The Iron and Steel Institute, 1947, vol. 157, Sept., pp. 81–88). The installation and development of recording and control instruments for use in a small open-hearth plant are described. Features of the system are the recording of the furnace-roof temperature and the gas and air flows on one chart, also an experimental automatic crown-temperature control. The instruments have proved to be of great value to the operators and have led to an increase in the melting rates.

The Question of Heat Transfer in a Molten Bath by Bubbles. M. A. Glinkov and V. S. Kocho. (Iron and Steel Institute, 1947, Translation Series, No. 325). This is an English translation of a paper which appeared in Bulletin de l'Académie des Sciences de l'U.R.S.S., Classes des Sciences Techniques, 1946, No. 10, pp. 1463–1472).—R. A. R.

Some Aspects of the Refining of High Phosphorus Iron. W. L. Kerlie. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 173–182). A comparison is made of the phosphorus and carbon equilibria and the factors affecting the rate of carbon removal in the refining of iron are explained on a thermodynamical basis. On the basis of the general equations developed, two examples of the refining of high-phosphorus iron are quoted and certain conclusions are reached from an examination of the results. The factors which determine the oxide requirements and processing time are considered and some notes on the more practical aspects of refining are given.

Use of Oxygen in Open-Hearth and Electric Furnaces. W. G. Gude. (Linde Air Products Co. : Foundry, 1947, vol. 75, Oct., pp. 74–77, 221–226). The equipment required for using oxygen in open-hearth and electric furnaces is described and operational experience is discussed. For carbon removal in the electric furnace an oxygen lance is pushed into the bath until the end of the tube is at the slag-metal interface. This method requires several 20-ft. tubes per heat. With jet injection the nozzle is kept 3 to 6 in. above the slag surface and it is pointed at the deepest portion of the bath. Tests to date have shown no increase in the consumption of electrodes. Trials with lances and jets in 30- and 70-ton electric furnaces indicate rates of carbon reduction of 0·5 points/min. for chromium steels and 1·0 points/min., or better, for unalloyed low-carbon steels.—R. A. R.

Practical Experience with the Use of Oxygen in Steelmaking. G. V. Slottman. (Journal of The Iron and Steel Institute, 1947, vol. 157, Nov., pp. 331–336). A general account is given of experience in the United States with the oxygenated-oil firing of open-hearth furnaces and the use of oxygen as a bath reagent. Some twenty furnaces of 60–200 tons capacity have been equipped for oxygenated-oil firing. The effect of oxygenation on output and fuel consumption varies with the condition of the furnace and the type of charge—the greatest improvements being obtained in small furnaces of the older types and with charges requiring long melt-down periods. The use of oxygen added directly to the bath through bare pipes for controlling the bath temperature and for shortening the refining period has become standard practice in several plants making low-carbon steel.

Production of High-Strength Steels with Improved Machining Characteristics. F. O. Johnson. (Metal Progress, 1947, vol. 52, Oct., pp. 565–567). A short

account is given of a process for improving the machinability of high-carbon steels by the addition of sodium sulphite in the ladle. Sufficient sulphite is added to increase the sulphur content of the steel to 0·04–0·06%. The sulphides formed are distributed throughout the metal, which increases machinability, and the sodium oxide reacts with any abrasive particles (such as silica and alumina) which may be present, to form a fluid slag. This sulphite treatment has very little effect on mechanical properties or heat-treatment characteristics—C. O.

30 Years' Advance in the Basic Openhearth. J. J. Golden and M. F. Yarotsky. (Metal Progress, 1947, vol. 52, Oct., pp. 572–577). A survey is made of the changes in the basic open-hearth practice of the Carnegie-Illinois Steel Corporation during the past thirty years. The topics discussed include furnace design, refractories, instrumentation, and the use of oxygen.—C. O.

Basic Open-Hearth Process. A. Barberot. (Chimie et Industrie, 1941, vol. 45, Jan., pp. 3–10; Feb., pp. 112–118). The history and theory of the basic open-hearth process are discussed, with particular reference to dephosphorization. Mention is also made of the Perrin-Ugine process—R. F. F.

The Basic Principles of the Desulphurization of Pig Iron and Steel. P. Bardenheuer and W. Geller. (Iron and Steel Institute, 1947, Translation Series, No. 324). This is an English translation of a paper which appeared in Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1934, vol. 16, No. 7, pp. 77–91 (*see* Journ. I. and S.I., 1934, No. I, p. 491).—R. A. R.

The Use of Sponge Iron in Swedish Steel Production. M. Tigerschiöld. (Jernkontorets Annaler, 1947, vol. 131, No. 9, pp. 295–339). [In Swedish]. The methods and costs of producing sponge iron in Sweden and experience gained in the production of steel from sponge iron is discussed. The account is divided into the following sections: (1) Predicting the composition of the sponge iron when using different raw materials; (2) costs of producing sponge iron; (3) previous practice in the smelting of sponge iron; (4) recent tests in the smelting of sponge iron and cost calculations for producing steel from it in basic arc furnaces, open-hearth furnaces, and high-frequency acid furnaces; (5) calculating and comparing the values of sponge iron and pig iron; and (6) selecting the method of production and the raw material for making sponge iron.—R. A. R.

Non-Ferrous Metals—Influence in Scrap for Steelmaking. E. R. Thews. (Iron and Steel, 1947, vol. 20, Oct., pp. 479–480). The influence of non-ferrous metals mechanically attached to steel scrap on open-hearth steelmaking and the effects on steelmaking practice and on the properties of the finished steel of zinc, lead, copper, nickel, chromium, manganese, and sulphur present as impurities, are discussed.—C. O.

Speeding Open Hearth Charging. R. J. Harry. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. 68–71). The author discusses the need for increasing the rate of charging of open-hearth furnaces which has resulted from the growing use of oxygen and light scrap in steelmaking. A comparison is made of the high and low types of charging machine. For speeding-up open-hearth charging the following recommendations are made: (1) Stock cars should be unloaded by high-speed magnetic cranes; (2) facilities should be provided for handling the increased production in the stockyard and for increasing the density of light scrap; (3) scrap boxes should be larger; (4) furnace doors should be larger; (5) the speed and capacity of charging machines should be increased.—C. O.

Ladle Deoxidation of Killed Steels with Silicon Carbide and Its Effect on Physical Properties and Hardenability. E. A. Loria and A. P. Thompson. (Transaction of American Society for Metals, 1947, vol. 39, pp. 651–665). A standard ladle-deoxidation practice using granular silicon carbide is described. It resulted in improved microstructure and physical properties of the steel.—R. A. R.

The Manufacture of Hollow Drill Steel in Australia. E. B. Suters. (B.H.P. Review, 1947, vol. 24, June, pp. 14–16). An account is given of the development of the manufacture of hollow-drill steel in Australia since 1938.—C. O.

Melting High-Speed Steel in the Basic Electric Arc Furnace. H. C. Bigge. (Iron Age, 1947, vol. 160, Oct. 16, pp. 118–136). A detailed description is given of the production of a heat of 18/4/1 tungsten–chromium–vanadium steel in a 7-ton basic electric-arc furnace. Graphs are given which show the correlation between metal and slag analyses and furnace additions. A series of coloured photographs illustrates the changes in the appearance of the slag as the heat progresses.—C. O.

FOUNDRY PRACTICE

Mass Production Methods Applied to the Manufacture of Cast Steel Bomb Bodies. P. H. Wilson. (Institute of British Foundrymen: Foundry Trade Journal, 1947, vol. 83, Oct. 23, pp. 149–155; Oct. 30, pp. 177–180; Nov. 6, pp. 197–203; Nov. 13, pp. 219–221). To meet the demands of the combined allied air forces during the war, a new British plant was constructed for the large-scale production of 500-lb. cast-steel bomb bodies. The steel for the bomb bodies was successfully produced in side-blown converters from cupola-melted steel scrap by a process used for the first time on a commercial scale; the method is described in detail. The castings were mass-produced in green-sand moulds in a highly mechanized steel foundry in which the various operations of sand-handling, core-making, moulding, stripping, cooling, fettling, testing, heat-treatment, shot-blasting, and internal grinding were integrated to give a steady production of 100 castings per hour.—J. R.

Iron Foundry Advances in the Chicago Area. K. H. Briggs. (Metal Progress, 1947, vol. 52, Oct., pp. 578–580). An outline is given of the technical development of iron-founding in the Chicago region during the last decade. The most outstanding advances have been the replacement of rule-of-thumb methods by scientific control, mechanization, improved melting methods, and inoculation.—C. O.

Wartime Plant Makes Large Steel Castings for Peacetime Uses. P. Dwyer. (Foundry, 1947, vol. 75, Sept., pp. 90–92, 212–214). An illustrated description is given of the New Castle, Pennsylvania, foundry plant of the United Engineering and Foundry Co., Pittsburgh. The foundry is equipped for the production of steel castings of the largest size.—C. O.

Plan for the Rebuilding of a Bombed Iron Foundry. (Fonderie, 1946, Feb., pp. 68–76). Plans for the reconstruction of a French iron foundry destroyed during the war are detailed. A list of plant and material required for equipping the foundry is given in an appendix.—J. C. R.

Experience Can Point Way to Better Castings. C. L. Frear. (Foundry, 1947, vol. 75, Oct., pp. 92–93, 206–220). A study is made, based on wartime experience of foundries making castings for ships, of methods of preventing the manufacture of faulty castings.—R. A. R.

Convection Currents in Gray Cast Iron. J. H. Schaum. (Foundry, 1947, vol. 75, Sept., pp. 66–69, 192–200). A detailed account is given of experiments made to

determine the rôle of convection currents in the solidification of grey-iron castings. Temperature measurements were made during the pouring of castings, by means of thermocouples in quartz tubes extending into the mould cavities, and the progress of solidification was studied macroscopically by bleeding castings. It was found that convection currents are of sufficient magnitude and duration in grey-iron castings, bottom-poured, with 370° F. superheat, to cause the hottest metal to flow from bottom to top in sections as small as 1¼ in. in dia., and as much as 10 in. high. Convection currents cease when the temperature of the metal falls below the liquidus. It is considered that if convection currents are sufficiently active in a casting to cause the top to become the hottest part, top risers should be used. —C. O.

Carbon Pick-Up in Melting Cast Iron. P. H. Shotton. (Institute of British Foundrymen: Foundry Trade Journal, 1947, vol. 83, Oct. 9, p. 115). Conditions affecting carbon pick-up by the metal during the melting of cast iron in the cupola, electric furnace, and rotary furnace are discussed. The factors favourable to increased carbon content in cupola practice are summarized as follows: (1) A long interval between taps, (2) the use of a coke of open texture, (3) a high coke bed, (4) insufficient air, and (5) a charge prone to carbon pick-up. —J. R.

The Use of Cupola Cokes Supersaturated with Moisture. G. Loose. (Ingénieur-Chimiste, 1946, vol. 28, Dec., pp. 89–110). An account is given of the combustion reactions that take place in the cupola and of the influence of the quality of the coke. The advantages of supersaturating poor-quality and small-size coke are discussed.—J. C. R.

Foundry Furnaces. (Overseas Engineer, 1947, vol. 21, Dec., pp. 158–164). Melting furnaces suitable for the small-scale manufacture of castings are described and illustrated.—C. O.

A Comparison of Moulds of Standard Composition and of Approximately Ingot Mould Sub-Committee Composition. W. L. Kerlie. (Journal of The Iron and Steel Institute, 1947, vol. 157, Nov., pp. 410–415). A number of ingot moulds representing two different compositions have been examined under carefully controlled conditions. Direct correlations between silicon content and mould life and phosphorus content and mould life were found, when all the moulds other than the abnormal ones were considered. The most important conclusion obtained was that with a mould composition close to the one recommended by the Ingot Mould Sub-Committee an average mould life of 130 casts is to be expected compared with an average of 100 for the standard composition.

Fourth Report of the Research Committee on High-Duty Cast Irons for General Engineering Purposes: Acicular Cast Irons. J. G. Pearce. (Institution of Mechanical Engineers, 1947, Preprint). An account is given of the composition, production, and properties of the acicular cast irons. These irons can be produced in any type of melting unit, and their mechanical properties, especially impact strength, are much superior to those of the best high-duty cast irons hitherto known.—C. O.

German Malleable Iron Foundries (With a Note on Schenck Fatigue Machines). British Intelligence Objectives Sub Committee, 1947, Final Report No. 1554: H.M. Stationery Office). Descriptions are given of the equipment and practice of German malleable-iron foundries varying in size from one with a capacity of less than 100 tons per month to a fully mechanized foundry with an output in excess of 1000 tons per month. A brief mention is made of two machines, manufactured by Carl Schenck Maschinenfabrik, Darmstadt, suitable for the fatigue-testing of large cast-iron specimens.—C. O.

The Annealability of White Iron in the Manufacture of Malleable Iron. S. W. Palmer. (Institute of British Foundrymen: Foundry Trade Journal, 1947, vol. 83, Oct. 2, pp. 87–94; Oct. 9, pp. 107–113; Oct. 16, pp. 129–135). An outline is given of research carried out and in progress at the British Cast Iron Research Association upon the annealing of white iron for the production of malleable iron. The work includes consideration of the influences of the manganese/sulphur ratio and the silicon content upon the graphitization, general microstructure, mechanical properties, and decarburization of whiteheart malleable iron, and also of the effects of different rates of heating and cooling during annealing.—J. R.

Annealing of Blackheart Malleable Cast Iron. M. Knotek. (Hutnické Listy, 1947, vol. 2, No. 2, pp. 31–36). [In Czech]. An improved method of heat-treatment for making blackheart malleable iron castings is described, and the effects of copper additions and of deoxidizing with aluminium are discussed.—R. A. R.

Cast Bolts for Pipe Joints. C. K. Donoho. (American Society of Mechanical Engineers: American Foundryman, 1947, vol. 12, Sept., pp. 24–27, 62). Descriptions are given of the manufacture in metal moulds and the properties of malleable cast-iron bolts and nuts to be used for joints in pipelines. The metal contains about 2·5% of carbon and 1–1·5% of copper. Details of corrosion tests on the bolts in 2% sulphuric acid are given. Austenitic malleable iron bolts containing manganese 1·25%, nickel 15·0%, copper 6%, and chromium 3·0% have also been made.—R. A. R.

Thirty Years in Steel Founding. H. W. Maack. (Metal Progress, 1947, vol. 52, Oct., pp. 631–635). A survey is made of progress during the last thirty years in the techniques and equipment used in the founding of steel valves and fittings.—C. O.

The Scope of Bond Clays in the Foundry. A. W. Ulmer. (Institute of Australian Foundrymen: Australasian Engineer, 1947, Sept. 8, pp. 39–48). A detailed survey is made of the properties and foundry applications of the clays. Particular attention is paid to the selection and preparation of natural and synthetic moulding sands and to cores.—C. O.

Reflections on Foundry Sands. M. Guédras. (Métallurgie, 1947, vol. 79, Apr., pp. 5–6). This is a brief general account of the properties of foundry sands and their reclamation.—J. C. R.

Sand Control. G. W. Anselman. (Chicago Regional Foundry Conference: Foundry, 1947, vol. 75, Oct., pp. 97, 164–172). The control of the grain-size of moulding sand and core sand, the proportions of clay and water to be added, and the properties of southern and western bentonites are discussed.—R. A. R.

Heat Absorption of Molding Sand. H. W. Dietert, E. J. Hasty, and R. L. Doelman. (Foundry, 1947, vol. 75, Sept., pp. 84–85, 228–232). The results of measurements of the thermal conductivity of bentonite-bonded silica sands are presented and discussed.—C. O.

Selecting Proper Pattern Equipment. R. Olsen. (New England Foundry Conference: Foundry, 1947, vol. 75, Sept., pp. 93, 215–216). Economic considerations involved in the selection of patterns for castings are briefly discussed.—C. O.

Molding Machines. E. A. Blake. (American Foundryman, 1947, vol. 12, Sept., pp. 31–35). Illustrated descriptions of American jolt-squeeze types of moulding machines are given, together with a table indicating the piston diameters and pressures required for different mould areas.—R. A. R.

Fundamentals of Molding Machines. E. A. Blake. (American Foundrymen's Association: Foundry, 1947, vol. 75, Sept., pp. 78–83). The applications of the common

types of moulding machine are discussed and illustrated.—C. O.

New Coreroom Is Model of Efficiency. P. Dwyer. (Foundry, 1947, vol. 75, Oct., pp. 66–71, 130). A detailed and illustrated description is given of the recently equipped core-making department of the Cleveland Foundry Co., Cleveland, Ohio.—R. A. R.

ROLLING-MILL PRACTICE

Calculation of Loads in Metal Strip Rolling. M. Cook and E. C. Larke. (Journal of the Institute of Metals, 1947, vol. 74, Oct., pp. 55–80). A method has been developed for computing rolling loads without the necessity for determining the influence of such factors as roll-face distortion. The method is based on the assumption that the magnitude of the pure work of rolling is independent of the number of passes used in effecting a given reduction in thickness. Good agreement has been found between calculated and experimentally determined loads for copper and steel over the range of reductions encountered in normal rolling-mill practice.—C. O.

Modern Small Rolling Mills. G. A. Phipps. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 247–261). Detailed descriptions with operating data are given of eight small rolling mills of American type installed in Great Britain for the production of rods, bars, angles, strip, and merchant sections. It is shown that this country is well supplied with rod and narrow strip mills running at low conversion costs and high shift outputs, but this cannot be said for merchant mills. Development of mills with finishing rolls 13 to 15 in. in dia. is long overdue.—R. A. R.

The Light-Section Mill at the Darlington Works of the Darlington and Simpson Rolling Mills, Ltd. (Journal of The Iron and Steel Institute, 1947, vol. 157, Nov., pp. 447–452). A description is given of the recently completed extensions to the mills of the Darlington and Simpson Rolling Mills, Ltd., which are designed to produce 75 tons/shift of standard window-frame and other light sections.—R. A. R.

Metallurgical Problems of Rails. G. Grenier. (Revue de l'Industrie Minérale, Comptes Rendus, 1947, Apr., pp. 219–225). This is a review of the different improvements made in the manufacture of rails and the tests carried out on rails.—R. F. F.

Homestead's 160 in. Plate Mill. W. H. Gilleland and W. D. Hacker. (Iron and Steel Engineer, 1947, vol. 24, Oct., pp. 35–43). An illustrated description is given of the principal features of a 160-in. plate mill recently installed at the Homestead, Pennsylvania, plant of the Carnegie-Illinois Steel Corporation. The main mill stand is a reversing four-high stand, with alloy-iron work rolls 38 in. in dia., and alloy-steel back-up rolls 59 in. in dia. The rolls are driven by two 5000-h.p., 700-V. D.C. reversing motors, through universal spindles 36 ft. long.—C. O.

Sheet and Tinplate Mills—Roll Dressing Formulæ. J. H. Mort. (Iron and Steel, 1947, vol. 20, July, pp. 351–356; Sept., pp. 431–435; Oct., pp. 481–482, 496). Formulæ for the dressing of rolls in orthodox two- and three-high mills are developed.—C. O.

Hot and Cold Strip Mills—Important Advances Made in the Chicago Area. E. D. Martin. (Metal Progress, 1947, vol. 52, Oct., pp. 561–564). The improvements in the technique of hot- and cold-rolling of steel strip brought about at the Chicago works of the Inland Steel Co. during the last thirty years are reviewed.

Manufacture of Aluminium Clad Steel Strip by Wickede Eisen und Stahlwerke. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 1567: H.M. Stationery Office). Details are given of the practice adopted by the Wickede Eisen und Stahlwerke for the production of "Feran"—an aluminium-clad steel strip. Deep-drawing-quality 3·5-mm. steel strip was cold-rolled between two 0·2-mm. strips of aluminium (containing 0·6–0·7% of silicon), the whole being reduced to 1·5 mm. in thickness in one pass.—C. O.

German Methods of Production of Aluminium Coated and Continuous Electroplated Steel Strip. (British Intelligence Objectives Sub-Committee, 1947, B.I.O.S. Final Report No. 1467: H.M. Stationery Office). Short accounts are given of the production in Germany of aluminium-coated steel strip, by cladding mild steel (less than 0·08% of carbon) with aluminium strip, and of zinc- and brass-plated steel strip, by continuous electroplating.—C. O.

Tungsten Carbide Rolls for Wire Flattening. E. C. Slick and R. E. White. (Iron Age, 1947, vol. 160, Oct. 9, pp. 74–77). The development of tungsten carbide rolls for flattening wire of very small section is described. Tungsten carbide sleeves with a 2½-in. face were press-fitted on to a steel sleeve shrunk on an arbor. Very great increases in roll life were obtained as compared with hardened steel rolls.—R. A. R.

Pilger Mills of the German Steel Tube Industry. (British Intelligence Objectives Sub-Committee, 1947. Final Report No. 1563: H.M. Stationery Office). Short descriptions are given of the thirteen Pilger mills operating in the British Zone of Germany. There was very little development work on the Pilger mill during the war, nor was there any great advance in the technique of mill operation.—C. O.

Mechanical and Electrical Features of Primary Hot-Rolling-Mill Auxiliaries. Section I.—Mechanical Features. W. W. Franklin. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 262–271). Designs of the following mechanical equipment for heavy hot rolling mills are described and discussed: (*a*) Roller tables; (*b*) breast rollers; (*c*) universal manipulators; (*d*) tilting fingers; (*e*) hydraulically operated manipulators; (*f*) screwdown gears; and (*g*) bloom shears.—R. A. R.

Mechanical and Electrical Features of Primary Hot-Rolling-Mill Auxiliaries. Section II.—Electrical Features. P. F. Grove. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 273–278). The application of electric drive to the machinery described in Section I of this paper (*see* preceeding abstract) is discussed together with other problems which arise in choosing and arranging suitable electrical apparatus to serve those who operate and maintain the mill and its auxiliary equipment.—R. A. R.

Reconditioning Hot Mill Rolls. R. S. Trimble. (Iron Age, 1947, vol. 160, Sept. 18, pp. 71–73; Sept. 25, pp. 77–80). This general review of modern techniques for reconditioning rolls contains discussion of surface quality and finish, grinding machines, and surface defects.—C. O.

The Electrical Drive of Rolling Mill Auxiliaries. S. A. Lewitt and H. Nielsen. (G.E.C. Journal, 1947, vol. 14, Aug., pp. 215–231). A detailed illustrated account is given of the electrical drives of the slabbing and hot and cold strip mills and continuous pickling plant at the Hawarden Bridge steelworks of Messrs. John Summers and Sons, Ltd., Shotton.—C. O.

Welding Applications in Ore Bridges. G. F. Wolfe. (Iron and Steel Engineer, 1947, vol. 24, Oct., pp. 84–85). A short account is given of the construction of a 15-ton capacity ore bridge in which all the components except the main truss and the cantilever spans were welded.—C. O.

Alloy Billet Grinder Features Oscillatory Head Movement. (Iron Age, 1947, vol. 160, Nov. 6, pp. 99–100).

A very brief description is given of an automatic machine for grinding the flat surfaces of stainless-steel slabs 30 ft. long × 24 in. wide × 4 in. thick.—C. O.

Maintenance of Diesel Switching Locomotives. R. W. van Saut, jun. (Iron and Steel Engineer, 1947, vol. 24, Sept., pp. 38–42). The effect of steelworks' conditions on the rate of wear of diesel shunting engines is discussed, and the main features of an efficient maintenance organization are outlined.—C. O.

Machines for Cleaning Strainers. G. Burkhardt. (Von Roll Mitteilungen, 1944, vol. 3, Dec., pp. 131–134). Descriptions are given of machines used for keeping clean the strainers and gratings covering cooling-water intakes.—R. A. R.

WELDING AND FLAME-CUTTING

Symposium on Metallurgy of Steel Welding. (British Welding Research Association, 1947). The object of this Symposium was to provide means of promoting the exchange of views on the work and programmes of the various committees of the British Welding Research Association. The Symposium was divided into four sections, the first two on weld metal, the third on hardened-zone cracking and viscosity of welding slags, and the fourth on current research. Abstracts of the papers follow.—R. A. R.

Constitution of Weld Metal—Results of Research Arising from Practical Problems. W. Andrews. (*Ibid.*, pp. 5–18). The research work on the constitution of weld metal, the results of which are given in this report, is covered by the three headings: (1) General chemical analysis of mild steel weld metal; (2) detailed examination by vacuum fusion and iodine extraction of selected weld deposits; and (3) examination of metallic-arc weld deposits made with anhydrous fluxes.—R. A. R.

A Preliminary Investigation of the Constitution of Mild-Steel Arc-Weld Deposits. H. A. Sloman, T. E. Rooney, and T. H. Schofield. (*Ibid.*, pp. 19–36). Descriptions are given of the methods used for determining the oxygen, hydrogen, and nitrogen contents of mild-steel weld deposits laid down under certain standard conditions, as well as the form in which the oxygen is present. The results obtained are presented and discussed. The effect of heat-treatment on the microstructure and on the precipitation of the nitride-bearing constituent, and the form of combination of sulphur as manganese sulphide were also studied.—R. A. R.

An Analytical Examination of Weld Deposits from Commercial Mild Steel Electrodes to Specification B.S.S. 639A and the Manufacture of Artificial Weld Metal. G. L. Hopkin. (*Ibid.*, pp. 37–39).

Comments on the Rôle of Hydrogen in Relation to the Cracking of Alloy Steels on Welding. G. L. Hopkin. (*Ibid.*, pp. 40–41).

The Relation between the Hydrogen Content of Weld Metal and Its Oxygen Content. L. Reeve. (*Ibid.*, pp. 48–54). The main purpose of the investigation reported was to determine the relationship between the FeO content of weld metal and its total and diffusible hydrogen content. It is shown that the total hydrogen content is slightly reduced and the diffusible hydrogen content considerably reduced when the FeO content in the weld is increased. An attempt is made to explain these results theoretically. The influence of micro-cavities and inclusions is also discussed with special reference to the influence of the active FeO content of the inclusions.—R. A. R.

Influence of Sulphur and Phosphorus on Weldability of Mild Steel. L. Reeve. (*Ibid.*, pp. 55–64). This paper reports certain observations on the influence of sulphur, phosphorus, and carbon on the cracking tendency of mild-steel weld metal, and the influence of the type of electrode on this phenomenon.—R. A. R.

Effect of Initial Heating Temperature on the Mechanical Properties of an Air-Hardened Ni-Cr-Mo Steel. J. A. Wheeler and V. Kondic. (*Ibid.*, pp. 71–83). An investigation is reported on the effect of the temperature to which an air-hardening steel is heated before quenching, on the mechanical properties of the resulting martensite. The temperature range covered is 850–1475° C., and it thus includes almost the whole range of heat-treatment temperatures characteristic of the hardened zone of a weld. The properties studied were ultimate tensile strength, elongation, reduction of area, and Vickers hardness.—R. A. R.

An Apparatus for the Determination of the Viscosity of Welding Slags. P. K. Gledhill. (*Ibid.*, pp. 88–93). A detailed description is given of an apparatus which was designed to measure the viscosity of welding slags at very high temperature. The rotating-cylinder method was employed, in which the inner cylinder is suspended from a torsion wire, and the logarithmic decrement of the damped oscillations is proportional to the viscosity.—R. A. R.

Electro-Physics of the Welding Arc. L. H. Orton. (*Ibid.*, pp. 98–100). An explanation is given of the work to be undertaken by the committee recently formed to study the electro-physics of the arc.—R. A. R.

Cracking of Welded Gas Mains. L. Reeve. (*Ibid.*, pp. 102–103). Work done and contemplated in establishing the causes of cracks near butt and fillet welds in gas mains and vessels associated with the treatment of crude town and coke-oven gas is briefly discussed. Tests in boiling nitrate solutions indicate that a minimum degree of stress must be present to produce failure. The most sensitive steels will fail at an imposed stress of less than 4 tons sq. in.—R. A. R.

Some Problems in the Approach to Welding Design. R. G. Braithwaite. (British Institute of Welding: Australasian Engineer, 1947, Aug. 7, pp. 45–49). It is shown that the design of efficient welded structures demands a more detailed study of the physical and metallurgical properties of metal than is required in riveted structures. The use of high-tensile steel in bridge construction is discussed, together with stress distribution in welds, and the factors involved in rigid joints.—R. A. R.

A Note on the Shrinkage and Distortion of Welded Joints. F. Guyot. (Arcos, 1946, No. 100, pp. 2357–2380; No. 101, pp. 2399–2404: Welding Journal, 1947, vol. 27, Sept., pp. 519-s–529-s). The author discusses theoretically the shrinkage parallel and perpendicular to butt- and fillet-welded joints. Practical applications of shrinkage curves which are derived are explained.—C.O.

PROPERTIES AND TESTS

The Ductility of Metals under General Conditions of Stress and Strain. J. E. Dorn and E. G. Thomsen. (Transactions of American Society for Metals, 1947, vol. 39, pp. 741–772). A method of calculating the finite strains that metals can undergo by combined stressing up to the instant of fracture, based on their flow and fracture characteristics, is described, and results obtained with it are compared with experimental data.—R. A. R.

A Semi-Quantitative Method for Measuring the Ductility of Chromium Electrodeposits. M. R. J. Wyllie. (Electrochemical Society, 1947, Preprint 92-5). A method is described by which the ability of uniformly

electrodeposited chromium plating on etched mild steel to deform can be estimated quantitatively. The cracks and discontinuities which penetrate to the base metal are estimated quantitatively, by an adaptation of the Dubpernell test, both before and after the specimen is elongated to an extension of approximately 1%.—C.O.

Measuring the Elastic and Plastic Properties of Wire. (Wire Industry, 1947, vol. 14, Nov., pp. 617–618). An adaption of the Hounsfield extensometer to measure strains of the order of 1/600,000 in./in. in wire is described. The use of the instrument for determining elastic and plastic deformations separately is indicated.—C. O.

Photoelasticity and Its Application. A. Pirard. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 11, pp. 537–545). The methods and apparatus used and some of the results obtained in the photo-elasticity laboratory of the University of Liége are described.—R. F. F.

Measurement of Young's Modulus at High Temperatures. M. H. Roberts and J. Nortcliffe. (Journal of The Iron and Steel Institute, 1947, vol. 157, Nov., pp. 345–348). Measurements have been made of Young's modulus of various steels from room temperature up to 1000° C. The method used was to cause a cylindrical rod of the steel to vibrate transversely with its fundamental frequency. From the frequency and the mass and dimensions of the bar, Young's modulus may be calculated. The results show a decrease in Young's modulus with temperature, the value at 600° C. being of the order of three-quarters of that at room temperature.

An X-Ray Method of Measuring Poisson's Ratio. R. F. Hanstock and E. H. Lloyd. (Proceedings of the Institution of Mechanical Engineers, 1947, vol. 157, War Emergency Issue No. 26, pp. 52–54). The measurement of Poisson's ratio in metals by an X-ray diffraction method is described. High-angle back-reflection X-ray photographs are taken of a small tensile test-piece, under various loads. At each load X-ray patterns are obtained for several angular settings of the specimen relative to the incident X-ray beam. The strains (as indicated by the linear shift of the diffraction lines) for any one load give a linear graph when plotted against a simple function of the orientation of the reflecting crystallographic planes. Poisson's ratio can be obtained directly from this graph. Experimental results for the alloy Hiduminium RR.56 are quoted.—C. O.

Dilatometry in the Range −50° to +150° C. P. Dubois and R. Walden. (Chimie et Industrie, 1942, vol. 47, May, pp. 540–542). Modifications to Chevenard's differential dilatometer, in order to obtain, in the range −50° to +150° C., dilatometric curves allowing the determination of expansion coefficients with an error of less than 2%, are described.—R. F. F.

The A. Dumez Type of Chronographic Recorder for the P. Chevenard Type of Mechanically Recording Differential Dilatometer. A. M. Dumez. (Bulletin du Cercle d'Études des Métaux, 1947, vol. 4, June, pp. 461–484). Details are given of linear and logarithmic chronographic recorders for use with Chevenard's mechanically recording differential dilatometer, and their applications.—J. C. R.

High-Magnification Stress-Strain Charts. J. O. Jones. (Machinery, 1947, vol. 71, Nov. 6, pp. 513–515). A short account is given of an electronic high-magnification recording extensometer developed by the Tinius Olsen Testing Machine Co., Philadelphia.—C. O.

Notched-Bead Slow-Bend Tests of Carbon–Manganese Steels. C. B. Voldrich, D. C. Martin, and O. E. Harder. (Welding Journal, 1947, vol. 27, Sept., pp. 489-s–506-s). The results are presented of notched-bead slow-bend tests on ½-in., ¾-in., and 1-in. plates of ten commercial rolled steels with different carbon and manganese contents, and also on ½-in. and 1-in. plates of three experimental rolled steels and three cast steels. The welds were made with three types of electrode and at three different rates of heat input. Notched slow-bend tests were made on unwelded specimens of all the steels. The main conclusions reached are: (1) Except for thin plates welded with a high rate of heat input, all the steels had a smaller bend angle in the welded than in the unwelded conditions; (2) the bend-angle ductility of welded specimens decreased as the carbon and manganese increased; (3) the bend angle decreased markedly as plate thickness increased; (4) the bend angle decreased as the heat input decreased; (5) the bend angles are the same for as-rolled and normalized plates; (6) the three types of electrode had the same influence on the bend angle; (7) the cast steels responded to welding substantially in the same way as rolled steels; (8) specimens welded with low-hydrogen electrodes had the same bend angle as similar specimens welded with high-hydrogen electrodes; and (9) there was a good correlation between the bend angle and maximum underbead hardness of similarly welded specimens.—C. O.

Bend Testing of Armature Binding Wire. (Steel, 1947, vol. 121, Oct. 27, p. 92). A simple device for carrying out continuous 180° bend tests over several feet of wire in order to detect intermittent defects is described.—C. O.

The Change in Mechanical Properties of Mild Steel under Repeated Impact. F. V. Warnock and J. A. Pope. (Proceedings of the Institution of Mechanical Engineers, 1947, vol. 157, War Emergency Issue No. 26, pp. 33–44). Experiments were carried out to investigate the variation under repeated impact of the impact yield strength and other mechanical properties of mild steel. Measurements were made with a scratch extensometer on specimens in a falling-tup type of tensile impact machine. It is concluded that: (1) The impact yield strength increases as the velocity of impact increases; (2) the impact yield strength decreases as a repeated-impact test proceeds; and (3) the impact yield strength of any given material varies greatly.—C. O.

Some Factors Affecting the Notched-Bar Impact Properties of Mild Steel. W. Barr and A. J. K. Honeyman. (Journal of The Iron and Steel Institute, 1947, vol. 157, Oct., pp. 243–246). Work carried out to confirm and extend conclusions stated previously by the authors on the influence of carbon and manganese on the notch sensitivity of mild steel is described. Steels were made to carefully controlled compositions and their structures and impact properties in the normalized and annealed states were determined. It was shown that increasing the carbon content raises the transition range and lowers the impact values of steels in the normalized condition, while increase of manganese content has the opposite effect. It was confirmed that an increased manganese/carbon ratio tends to give a steel with a finer McQuaid-Ehn grain-size and a finer structure in the normalized condition. A comparison of steels made with and without a grain-controlling addition of aluminium showed that the manganese/carbon ratio may be as effective as grain-size control in reducing notch sensitivity. A study of the relationship between the notched-bar impact value and the degree of cleavage in the fracture showed that the latter is not the best criterion by which to judge notch sensitivity. Material which gives a predominating cleavage fracture will be satisfactory in service, provided that an appreciable amount of energy is absorbed before a crack is initiated.

The Cold Work Hardening Properties of Stainless Steel in Compression. F. K. Bloom, G. N. Goller, and P. G. Mabus. (Transactions of American Society for Metals, 1947, vol. 39, pp. 843–864). The cold-work-

hardening properties of chromium–nickel and chromium stainless steels were studied by a special compression test. Increasing the nickel content consistently decreases the cold-work-hardening of chromium–nickel alloys, whilst the effect of chromium depends on the nickel content.—R. A. R.

Strain Gauging—Investigation of Behaviour of Materials in Service. (Iron and Steel, 1947, vol. 20, Oct., pp. 492–493 : British Steelmaker, 1947, vol. 13, Oct., pp. 528–530). A short description is given of the use of the electrical strain gauge for studying the behaviour of engineering components in service.—C. O.

Failures in Guided Bend Qualification Test Often Due to High-Tensile Pipe. L. K. Stringham. (Welding Journal, 1947, vol. 27, Sept., pp. 784–785). An investigation of failures in guided-bend tests of welds made by welders in qualification tests has shown that the cause is often the abnormally high tensile strength of the steel used. —C. O.

The Effect of Composition on the Fatigue Strength of Decarburized Steel. L. R. Jackson and T. E. Pochapsky. (Transactions of American Society for Metals, 1947, vol. 39, pp. 45–57). The fatigue strengths of nickel steel and chromium–molybdenum steel were investigated, and it was found that : (1) The core hardness of decarburized fatigue-test specimens does not have as great an influence on the fatigue strength as the hardness of the decarburized skin ; (2) the damaging effect of decarburization is not so pronounced at the high-stress end of the fatigue curve as at the endurance limit ; and (3) the fatigue strength of decarburized steels is primarily dependent on the strength of the ferrite in the relatively " carbon-free " decarburized zone.—R. A. R.

METALLOGRAPHY

Decarburization during Annealing of Malleable Iron. H. A. Schwartz and J. Hedberg. (Transactions of American Society for Metals, 1947, vol. 39, pp. 61–70). Metallographic investigations of the rims formed on white iron at various temperatures and in various environments, and chemical analyses of inclusions were carried out. White cast iron can be substantially decarburized even above the A_3 point of silicoferrite. The formation of grain-boundary inclusions, which have often been called FeO grain boundaries, occurs only in the presence of silicon, and these compounds are, therefore, probably silicates.—R. A. R.

Structural Variations in Gas Turbine Alloys Revealed by the Stress Rupture Test. N. J. Grant. (Transactions of American Society for Metals, 1947, vol. 39, pp. 335–359). In order to determine why the properties of some casts of heat-resisting high-strength alloys of the cobalt–chromium–molybdenum–titanium system differed from those of other casts the effects of preheating the mould and of the casting temperature were studied. It was found that although coarser-grained structures give higher strength values, a certain degree of coarseness must not be exceeded because grain orientation plays too large a rôle, causing early failure in some cases and extremely long times to rupture in others.—R. A. R.

The Effect of Composition and Structural Changes on the Rupture Properties of Certain Heat Resistant Alloys at 1500° F. N. J. Grant. (Transactions of American Society for Metals, 1947, vol. 39, pp. 368–402). A 69/23/6 cobalt–chromium–molybdenum alloy (vitallium) and a 32/32/26/6 cobalt–nickel–chromium–molybdenum alloy (No. 6059) were tested to study the effect of the casting conditions and microstructure on the mechanical properties. The effects previously reported (*see* preceding abstract) were confirmed. An addition of 1% of manganese makes the alloys stronger and improves the ductility. —R. A. R.

Changes in Austenitic Chromium–Nickel Steels during Exposures at 1100° to 1700° F. P. Payson and C. H. Savage. (Transactions of American Society for Metals, 1947, vol. 39, pp. 404–439). *See* Journ. I and S.I., 1946, No. II, p. 140A.—R. A. R.

Transformations in Krupp-Type Carburizing Steels. A. R. Troiano and J. E. DeMoss. (Transactions of American Society for Metals, 1947, vol. 39, pp. 788–798). *See* Journ. I. and S.I., 1946, No. II, p. 141A.—R. A. R.

Stability of Austenite in Stainless Steels. C. B. Post and W. S. Eberly. (Transactions of American Society for Metals, 1947, vol. 39, pp. 868–888). *See* Journ. I. and S.I., 1946, No. II, p. 140A.—R. A. R.

Isothermal Transformation of Austenite. A. Hultgren. (Transactions of American Society for Metals, 1947, vol. 39, pp. 915–989). *See* Journ. I. and S.I., 1946, No. II, p. 141A.—R. A. R.

Austenite Transformation Above and Within the Martensite Range. R. T. Howard, jun., and M. Cohen. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2283 : Metals Technology, 1947, vol. 14, Sept.). The results are presented of a quantitative study of the austenite → martensite transformation in five high-carbon and nickel steels, using a modified form of the Greninger-Troiano metallographic technique. Measurements of the isothermal transformations at temperatures above and below the M_s point have enabled the lower part of the conventional austenite transformation diagram to be drawn in some detail.—C. O.

Study of the Transformations of Annealed Carbon Steels with the Aid of the Iron–Carbon Diagram. C. Chaussin. (Métallurgie, 1947, vol. 79, Jan., pp. 15–17). After a general description of the iron–carbon diagram and the changing structure of steel as it cools and solidifies, an account is given of the changes that take place on annealing.—J. C. R.

The Nomenclature of the Decomposition Products of Austenite. H. Jolivet. (Bulletin du Cercle d'Études des Métaux, 1942, vol. 3, Dec., pp. 241–263). After a review of the origin of the terms austenite, martensite, troostite, osmondite, sorbite, and pearlite, the author considers the basis on which a classification might be formed in which constituents resulting from the same mode of decomposition would be grouped together and then subdivided, according to the form, colour, and disposition of the elements in the structure. Particular reference is made to the structures in the Ar′, Ar″ and Ar‴ transformation ranges.—J. C. R.

Note on the Graphic Determination of the Ac_3 Point by the Dilatometric Method. J. Paillard. (Bulletin du Cercle d'Études des Métaux, 1945, vol. 4, June, pp. 205–211).—J. C. R.

CORROSION

Ferric Iron Corrosion During Acid Cleaning. F. N. Alquist, J. L. Wasco, and A. A. Robinson. (Corrosion, 1947, vol. 3, Oct., pp. 482–487). A short account is given of the mechanism of the corrosion of heat-exchanger units during the removal of ferric oxide deposits by acid. This corrosion, which is caused by the reduction of the ferric ions formed during the cleaning, can be effectively prevented by the addition of glyoxal to the cleaning acids.—C. O.

Principles of Immersion and Humidity Testing of Metal Protective Paints. A. C. Elm. (A.S.T.M. Bulletin, 1946, vol. 142, Oct., pp. 9–27 : Corrosion, 1947, vol. 3, Oct., pp. 501–509). The author considers that in the design of testing methods there are three steps : (1) The formulation of a theory of what is needed ; (2) the development of a testing instrument and/or technique ; and (3) the extensive use of the instrument or technique to show what relationship it has with actual service conditions. Each of these steps is discussed with reference to tests in humid atmospheres or aqueous solutions of the protection afforded by paints on metal.—C. O.

Use of Magnesium, Zinc, Aluminium, and Their Alloys in the Cathodic Protection of Steel in Salt Water. R. R. Rogers and C. E. Viens. (Canadian Metals and Metallurgical Industries, 1947, vol. 10, Sept., pp. 16–18, 36–38). An investigation of the protection against corrosion afforded to steel immersed in salt water by certain non-ferrous metals and alloys is reported. Pure magnesium and high-magnesium alloys were found to give very good protection, although the anode corrosion rate was high. Pure zinc gave fairly good protection, with a low anode corrosion rate, while pure aluminium afforded very little protection.—C. O.

Results of Long Term Atmospheric Corrosion Tests Carried Out in Belgium on Steels Protected by Painting. J. Lagasse. (Ossature Métallique, 1947, vol. 12, Nov., pp. 482–488). Details are given of final examinations of steel plates protected by different paints and exposed to the atmosphere in various districts of Belgium for periods of three, five, and eight years.—J. C. R.

Apparatus for Testing Sea-Mist Corrosion in a Large Test-Chamber. J. Lucas and B. Colombet. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 194–200). The construction, operation, and uses of this apparatus are described. The corrosion obtained on the test-pieces is characterized by a close network of fine pits over the whole surface. On ferrous metals the rust coatings are very adherent. Test-pieces of all kinds can be tested simultaneously.—R. F. F.

Anti-Corrosion Surface Treatments and Protective Coatings. M. Ballay. (Chimie et Industrie, 1941, vol. 46, Aug., pp. 133–146). *See* Journ. I. and S.I., 1947, vol. 157, Dec., p. 642.

Effect of Carbide Structure on the Corrosion Resistance of Steel. R. W. Manuel. (Corrosion, 1947, vol. 3, Sept., pp. 415–431 : Steel, 1947, vol. 121, Oct. 13, pp. 82–84). A short survey is made of the literature on the effect of carbide structure on the corrosion of steel. Several examples of the microstructures which have been found associated with corrosion in upset or welded steel pipe-lines and oil-well tubing are illustrated. In general, the carbide structure appears to influence corrosion resistance under conditions in which solid corrosion products form on the surface. Well-formed pearlite with straight and continuous lamellæ is more corrosion resistant than spheroidized pearlite. It is considered that this fact may be due to an " anchoring " action of the continuous lamellae, or to a reinforcing action on the film of corrosion products.—C. O.

Use of Dehydration in Combating Internal Corrosion in Products Pipeline Systems. H. K. Phipps. (American Petroleum Institute : Corrosion, 1947, vol. 3, Sept., pp. 458–465). A method for removing free water from petroleum products, by passing them through beds of activated alumina, is described. Since dehydration was adopted on a 176-mile stretch of 6-in. dia. oil pipeline in Kansas, no corrosion has been experienced.—C. O.

Quantitative Evaluation of Intergranular Corrosion of 18-8 Ti. F. J. Phillips. (Transactions of American Society for Metals, 1947, vol. 39, pp. 891–906). *See* Journ. I. and S.I., 1947, vol. 155, p. 159.—R. A. R.

An Attempt to Select a Suitable Specimen for the Study of Corrosion Cracking in 18-8 Steel. M. H. Springer, E. V. Succop, D. S. McKinney, and M. A. Scheil. (Welding Journal, 1947, vol. 27, Sept., pp. 530-s–538-s). Tests of the resistance to corrosion by magnesium chloride solutions of specimens of chromium-nickel stainless steels, subjected to various amounts of plastic deformation and stress-relieved at different temperatures, are described. It was found that for corrosion cracking to occur, the metal must be subjected to a certain minimum tensile stress, the magnitude of which may be dependent on the heat-treatment and the corroding environment. —C. O.

Passivity in Chromium-Iron Alloys ; Absorbed Iron Films on Chromium. H. H. Uhlig. (American Institute of Mining and Metallurgical Engineers, Technical Publication, No. 2243 : Metals Technology, 1947, vol. 14, Sept.). Experiments are described which were made to determine whether iron could be made passive by mere physical contact with chromium. When thin layers of iron produced on chromium by electrodeposition or evaporation were immersed in nitric acid, it was found that the iron at the interface was always passive. The results confirm the electron-configuration theory of passivity.—C. O.

Is Cast Iron Superior in Corrosion Resistance to Steel ? R. W. White. (Materials and Methods, 1947, vol. 26, Aug., pp. 82–85). A description is given of an experimental comparison of the corrosion resistance of hot-rolled steel and cast iron, both uncoated and with organic coatings. The steel was found to be superior because the shot-blasted cast-iron surface offered a greater area to corrosion and led to a lack of uniformity of the protective coatings.—C. O.

ANALYSIS

Spot Tests for Steel. Cr, Ni, Si, Mn. W. E. Thrun and C. H. Bartelt. (Iron Age, 1947, vol. 160, Oct. 23, pp. 40–42). Data for the rapid and approximate estimation of chromium, nickel, silicon, and manganese in steels by the comparison of the colour intensities of spots on impregnated filter paper are described.—R. A. R.

Observations on a Method for the Potentiometric Determination of Cobalt in Steel. C. Long and R. Meyer. (Chimie Analytique, 1945, vol. 27, Feb., pp. 27–28). The method discussed is one previously described by Dickens and Maassen. It depends on attacking the steel with a mixture of hydrochloric, nitric, and perchloric acids, and the points emphasized are the necessity of standardizing the solutions on the basis of the electrolytic determination of cobalt nitrate and the importance, when attacking steels containing chromium, of keeping the quantity of the mixture of hydrochloric, nitric, and perchloric acids below a certain maximum value.—M. A. V.

Spectrographic Analysis of Steel. S. C. Baker. (Journal and Proceedings of the Royal Society of New South Wales, 1947, vol. 80, pp. 227–236). A description is given of an investigation of the technique for the spectrographic analysis of open-hearth steels by the standard 15,000-V. condensed-spark method. Particular attention was paid to the adjustment of the spectrograph and microphotometer and the processing of the plates. In conclusion the author gives an outline of the analysis of a steel by the " arcing to completion " method.—C. O.

Spectrochemical Analysis of High Copper in Cast Iron and Steel. W. R. Kennedy. (Journal of the Optical Society of America, 1947, vol. 37, Mar., pp. 154–158). The difficulties of the spectrochemical analysis of high percentages of copper in ferrous materials are explained. Details are given of a method for determining 0·15–0·80% of copper in steel or cast iron, using a spark, and of a

method for 0·50–2·50% of copper in cast iron, using an A.C. arc.—C. O.

Employment of Spectrum Analysis in Metallurgical Industry. A. K. Pokorný. (Hutnické Listy, 1947, vol. 2, No. 3, pp. 51–57). [In Czech]. Methods of applying spectrum analysis for testing ores and controlling melting processes are described.—R. A. R.

Spectrographic Determination of Boron in Steels. O. Masi. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 205–212). The spectrographic analysis of boron in plain carbon steels was carried out by sparking between two electrodes coated with a solution of the sample to be examined. The range of boron concentrations studied was 0·03–2·01%. Although this method was not sensitive enough for low percentages, for contents over 0·2–0·3% the variations in the determinations were of the order of 15%. The study of the variations in intensity between the iron and boron lines, under different experimental conditions led to the conclusion that there is probably a process of differential volatilization between iron and boron, and consequently, a variation in intensity with time and the discharge zone.—R. A. R.

Qualitative and Quantitative Spectroanalysis. M. Choul. (Métallurgie, 1947, vol. 79, Feb., pp. 17–20). A brief account is given of qualitative and quantitative methods of spectrographic analysis.—J. C .R.

Spectrographic Analysis of Low-Alloy Steel—A Statistical Examination of Sources of Error. H. T. Shirley, E. Elliott, and Joyce Meeds. (Journal of The Iron and Steel Institute, 1947, vol. 157, Nov., pp. 391–409). The main study of the sources of error in spectrographic analysis was made with a single sample of steel (C 0·30%, Si 0·25%, Mn 0·58%, Cr 0·77%, Ni 2·75%, Mn 0·50%). Readings were taken for silicon, manganese, chromium, nickel, and molybdenum from 30 spectrograms from each of thirty-one plates. These plates were taken under conditions involving excitation by means of normal Hilger equipment consisting of a 15,000-V. transformer, a 0·005-μF. condenser, and no added inductance.

The results of some 60,000 readings from upwards of 9000 individual lines have been treated statistically to obtain estimates of variability contributions from the three main sources: (1) Excitation response, (2) small-scale plate variability, (3) microphotometry. The greatest variability in excitation response occurred when using two pencil electrodes cut from the steel sample. Better results were obtained with either silver or graphite auxiliary electrodes. The standard deviation for excitation variability then corresponded to 1·5% of the chromium content, when using a chromium (II) line with an iron (I) line; but when using a new line-pair, composed of two singly ionized lines, this was reduced to 0·9%. Small-scale plate variability was studied by measurement of two independent 0·5-mm. lengths from each line, and Ilford thin-film half-tone plates were outstandingly good in this connection, with a standard deviation corresponding to about 0·7% of the chromium content. The microphotometry error was of the order of 1% of the chromium content. Total variability from these three sources, therefore, corresponded to a standard deviation of the order of 2% for the chromium (II)/iron (I) line-pair. By using the chromium (II)/iron (II) line-pair, this was reduced to 1·9%, the considerable improvement in sparking response being offset to some extent by lower sensitivity of this line-pair.

Results published by Vincent and Sawyer indicate a substantially lower overall variability, a figure of 0·7% of the content being quoted for 1% of chromium. It is shown that, in comparison with the results from the present investigation for the new chromium–iron line-pair, this lower overall variability is in great part due to a very small photometry error in the American work, the figure quoted corresponding, under the present authors' conditions, to a standard deviation of only 0·25 mm. in reading an individual line with a 50-cm. full-scale deflection.

Remarks on Spectrographs. P. Coheur. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 11, pp. 546–548). Grating spectrographs are compared to "Littrow" prismatic spectrographs with special reference to their resolving and dispersing powers.—R. F. F.

Mobile Laboratory Speeds Steel Analysis. H. A. Tuttle and G. A. Nahstoll. (Iron Age, 1947, vol. 160, Sept. 25, pp. 68–72). A detailed description is given of a mobile laboratory designed at the Rouge works of the Ford Motor Company for the rapid determination of the type of steel scrap, bar stock, etc. The equipment consists of a Spekker Steeloscope and chemical spot-test apparatus, mounted on a van chassis.—C. O.

Modifications to the Method of Determining Carbon in Steels and Ferrous Alloys by Combustion in Oxygen. C. Long. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 190–193). The quantitative study of factors in the combustion method has shown the importance of the rate of flow of the oxygen. This study has led to the introduction of a number of modifications to the method of determining carbon, in order to obtain a rapid and complete combustion. This modified method permits the determination of alloys with less than 0·050% with a margin of error of less than ± 0·005%.—R. F. F.

Use of Photoelectric Methods in Volumetric Analysis. A. Ringbom. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 304–308). The application of photoelectric methods to colorimetric titration is described. The determination of the *p*H can be carried out with an error of less than ± 0·01 *p*H units. The possibility of measuring oxidation-reduction potentials by means of indicators is considered, and theoretical considerations are applied to the titration of ferrous iron with a solution of ceric sulphate, using the ferrous complex of *o*-phenanthroline as indicator.—R. F. F.

A New Apparatus for the Colorimetric Determination of *p*H and the Comparison of Colours: The Compensating Chromometer. J. Rourillois. (Chimie et Industrie, 1941, vol. 45, Mar., Special Issue, pp. 309–312).—R. F. F.

Determining the Calorific Value of Coals and Coke, without the Use of Calorimetric Measurements. H. Lefebvre and C. Georgiadis. (Chimie et Industrie, 1941, vol. 46, Aug., pp. 147–158). After having analysed 1200 samples of coals and an equal number of cokes, and having carefully measured, with a Mahler bomb and a water calorimeter, their calorific value, the authors evolved two formulæ for the calculation of the calorific value of a coal or coke, when its humidity, ash, and volatile-matter contents have been determined. This calculated calorific value differs from that obtained by the best direct determination by only a few calories, often less than 15, hardly ever more than 35, in the range 7000–8000 cal./g.—R. F. R.

An Electrostatic Method for Determining Dust in Wet Gases. E. A. K. Patrick. (Journal of the Society of Chemical Industry, 1947, vol. 66, Aug., pp. 249–253). A small electrostatic precipitator is described which enables determinations to be made of the dust content of gases, including those which are inflammable and have dew points above room temperature, the majority of the dust particles being below 1 micron in dia.—R. A. R.

Determination of the Ash Content in Solid Fuels. H. Moureu, A. Marie de Ficquelmont, and R. Sabourin. (Annales des Mines et des Carburants, 1943, Series 14, vol. 3, pp. 209–227). Results are presented of tests carried out on 29 solid fuels, including charcoal, peat, brown coal, anthracite, and coke, to determine their ash content. Best results were obtained when the combustion temperature was between 800°C. and 850°C., and within this range the ash content was independent of the temperature and the period of heating if the latter were not less than 2 hr.—J. C. R.

ECONOMICS AND STATISTICS

Indian Iron and Steel—Recent Developments in the Industry. J. S. Vatchagandhy. (Iron and Steel, 1947, vol. 20, Oct., p. 485–491). A detailed review is made of the expansion during the last two decades in the iron- and steel-making capacity of the principal Indian works.—C. O.

The Iron and Steel Industry of Sweden. (British Iron and Steel Federation, Monthly Statistical Bulletin, 1947, vol. 22, Oct., pp. 1–38). The Swedish iron industry grew up in Central Sweden, which at one time was responsible for 40% of the world's iron production. In later years developments in other countries have limited Swedish exports of iron and steel to highly specialized products and Sweden is now an importer of steel. Comparative figures of capacity and output are given for the years 1938 and 1945. Doubts have been expressed as to the ability of Sweden to continue making charcoal pig iron, and it is suggested that sponge iron may take its place. In the interests of self-sufficiency, developments have taken place in the North Swedish province of Norrland where the Norrbottens Järnverks A/B. has been set up as a state-owned plant. This has the advantages of proximity to the Lapland orefields and of cheap local hydro-electric power. Further development of this plant is envisaged. It is considered that short-term prospects for United Kingdom exports of iron and steel to Sweden are favourable in view of Sweden's accumulated sterling balance, but the long-term position is not certain.—M. A. V.

Extract from Järnverksföreningens Statistics for the First and Second Quarters of 1947. (Jernkontorets Annaler, 1947, vol. 131, No. 10, pp. 487–492). [In Swedish]. Tables showing the Swedish production, consumption, exports, imports, and furnaces, relative to the manufacture of pig iron, ferro-alloys, steel ingots, castings, and rolled products, etc. are given.—R. A. R.

Blast Furnaces, Open-Hearth and Electric Furnaces in France. A. Witzig. (Echo des Mines, 1947, Oct., p. 163). A list of blast-furnaces and open-hearth furnaces in existence and in operation is given, with statistics for the production of iron and steel in France. Some data on electric-furnace steel production are included.—R. F. F.

The Iron and Steel Industry of Italy. (Echo des Mines, 1947, Oct., p. 163). The organization and productive capacity of the iron and steel industry in Italy are described.—R. F. F.

German Industrial Capital Equipment Available as Reparations. (Board of Trade Journal, 1947, vol. 153, Nov. 15, pp. 1984–1993). The plants notified by the British and American Commanders as available for reparations in their respective zones in Germany are listed. These include plants directly engaged in the production of war material and other surplus blast-furnace and steel plants. Separate lists are given for North-Rhine/Westphalia, Lower Saxony, Schleswig-Holstein, Hamburg, Bavaria, Hesse, Württemberg-Baden, Bremen, and the Berlin district.—M. A. V.

Modernisation of the Iron and Steel Industry. (4) Richard Johnson and Nephew, and Steel, Peech and Tozer. (Man and Metal, 1947, vol. 24, Nov., pp. 168–169). The continuous rod mill at Messrs. Richard Johnson and Nephew's Bradford Works is being reconstructed. The new installation will be a straight-through mill, and the output will be approximately doubled. On the completion of this, the wire mill will probably be reorganized with an increase of galvanizing capacity of about 30%. Messrs. Steel, Peech and Tozer have installed a new cogging mill and are reconstructing steel furnaces with a view to an eventual increase in capacity of 20%. This reconstruction includes conversion to oil-firing. The wheel mill is being rebuilt and a 30% increase of output is hoped for. The new plant includes a 6000-ton forging press, and hydro-electric power will be used instead of steam power as at present. Various other projects are in hand to save fuel.—M. A. V.

European Four-Year Programme for Coal, Iron and Steel. (Iron and Coal Trades Review, 1947, vol. 155, Nov. 7, pp. 893–897). A four-year programme for the European coal, iron, and steel industries is outlined in extracts taken from Reports drawn up by the Technical Committees of the Conference of European Economic Co-operation, in connection with the plan for American assistance to Europe.—J. R.

MISCELLANEOUS

Graphical Correlation—Its Application to Steel Mill Problems. C. R. Taylor. (Iron Age, 1947, vol. 160, Nov. 6, pp. 78–84). The use of the method of graphical correlation for determining the relationship between variables is illustrated by an example in which the effect of temperature, time between ferrosilicon addition and tapping, and amount of ferrosilicon addition, on the rate of deep-etching of a steel were studied.—C. O.

Past and Present Developments in Oxygen Generation. E. P. Stevenson. (American Philosophical Society: Steel, 1947, vol. 121, Oct. 20, pp. 114–118, 129; Oct. 27, pp. 98–102). A short account is given of the history of the separation of oxygen from the air by liquefaction. The principles of the low-pressure reversing-heat-exchanger method, which is in common use today, are discussed.—C. O.

Storage Yard Materials Handling. F. C. Wier. (Iron and Steel Engineer, 1947, vol. 24, Oct., pp. 51–57). The layout and operation of a yard for the storage and classification of raw materials, expecially scrap and finished products, for two steelworks of the Timken Roller Bearing Co., Canton, Ohio, are described and illustrated.—C. O.

Handling Materials for Steel Making. C. M. Parker. (Conference on Materials Handling: Industrial Heating, 1947, vol. 14, Aug., pp. 1296–1300). A general account is given of the magnitude of the material-handling problem of the iron and steel industry.—C. O.

"Schwingmetall"—A Process for Bonding Rubber to Steel Used Primarily for Mountings to Eliminate Vibrations. (British Intelligence Objectives Sub-Committee, 1947, Miscellaneous Report No. 40: H.M. Stationery Office). A short account is given of the "Schwingmetall" process for the elimination of vibration in machinery mountings. Soft rubber is bonded by a cement between two pieces of metal (usually steel or brass-plated steel), one of which is attached to the foundation and the other to the machinery. Details are given of the compositions of the rubber used.—C. O.

Characteristics of Japanese Naval Vessels—Gas Turbines. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./1150: H.M. Stationery Office).

BOOK NOTICES and NEW PUBLICATIONS

BOOK NOTICES

MOND NICKEL COMPANY, LTD. "*Nickel Cast Iron.*" [Loose Leaf]. London, 1947 : The Company.

This handbook is a revised and amplified edition of a similar book, first produced in 1939. Its object is to provide, in a convenient form, data covering all aspects of the production, properties, and applications of the many types of nickel cast irons in use today. Previously published information is collected and classified, but is supplemented by general instructions and practical notes on the production of the different types of nickel cast iron in the foundry, including such important points as choice of raw materials, furnace operations, methods of alloying, notes on moulding practice, and heat-treatment. Some of the later sections of the book deal with certain laboratory aspects, including the chemical analysis of nickel cast irons, and typical microstructures. A series of data sheets are also included, covering recommended compositions for applications in a number of specified fields.—R. E.

VAN SOMEREN, E. H. S. "*Spectrochemical Abstracts,*" Vol. III, 1940–1945. 8vo, pp. 112. London : Adam Hilger, Ltd.

This volume follows Vol. I, covering 1933–37, and Vol. II, covering 1938–39 ; consequently the abstracts are numbered consecutively with those in the earlier volumes. Part I of the present volume consists of authors and references arranged alphabetically under the authors' names, an index to elements in non-metallic substances, and an index to elements in metals. Part II consists of abstracts arranged according to substances analysed, apparatus, methods, and light-source theory.—R. E.

BRITISH COMPRESSED AIR SOCIETY. "*B.C.A.S. Handbook on Pneumatic Equipment.*" 8vo, pp. vi + 191. Illustrated. London : 1947. The Society. (Price 20*s.*).

This book will enable potential users of pneumatic equipment to recognize and appreciate the various types of air compressors, pneumatic tools, and ancillary equipment available ; it gives an adequate list of British manufacturers capable of supplying the plant described. The various industries in which compressed air is used are also enumerated.—R. E.

DE SY, A. "*Siderurgie en Electrosiderurgie. Band 1*" (Leerboek der Algemene Metallurgie, Siderurgie, Electrosiderurgie en Metallografie, Deel II). 8vo, pp. 454. Illustrated. Antwerpen, 1947 : Uitgeversmij. N.V. Standaard-Boekhandel.

This is the second of a series of volumes on metallurgy, and is devoted to methods for the manufacture of pig iron and steel. The book opens with a brief account of the history of ironmaking and the extent of the steelmaking industry. Particulars are included of the steelmaking plants in Belgium, and the number of blast-furnaces, converters, open-hearth furnaces, electric furnaces, coke-ovens, and rolling-mills operated at the individual plants are shown. In Part I the author discusses the raw materials for the manufacture of pig iron, the design and operation of the blast-furnace, hot-blast stoves, and gas cleaning. Part II is concerned with direct methods of steel manufacture, the open-hearth process, and refining in the converter. Particulars are also included of the Perrin process.—R. E.

NEW PUBLICATIONS

AMERICAN SOCIETY FOR METALS. "*Introductory Physical Metallurgy.*" By C. W. Mason. A Series of Lectures on Introductory Physical Metallurgy presented to members of the A.S.M. during the Twenty-Ninth Metal Congress and Exposition, Chicago, Oct. 18 to 24, 1947. 8vo, pp. 134. Illustrated. Cleveland, Ohio : The Society.

BRITISH COMPRESSED AIR SOCIETY. "*B.C.A.S. Handbook on Pneumatic Equipment.*" 8vo, pp. vi + 191. Illustrated. London, 1947 : The Society. (Price 20*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 309 : 1947. "*Whiteheart Malleable Iron Castings,*" 8vo, pp. 11. London : The Institution. (Price 2*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 310 : 1947. "*Blackheart Malleable Iron Castings.*" 8vo, pp. 9. London : The Institution. (Price 2*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 1313 : 1947. "*Fraction-Defective Charts for Quality Control.*" 8vo, pp. 40. London : The Institution. (Price 6*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 1407 : 1947. "*High Carbon Bright Steel (Silver Steel).*" 8vo, pp. 7. London : The Institution. (Price 2*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 1408 : 1947. "*Hard Drawn Steel Wire for Springs.*" 8vo, pp. 15. London : The Institution. (Price 2*s.*).

BROWN, G. H., C. N. HOYLER, and R. A. BIERWIRTH. "*Theory and Application of Radio Frequency Heating.*" Pp. 370. New York, 1947 : D. Van Nostrand Co., Inc. (Price $6.50).

Das Giessereiwesen in gemeinfasslicher Darstellung. Hrsg. von der Wirtschaftsgruppe Giesserei-Industrie. 2. Aufl. Düsseldorf und Possneck, 1945 : Giesserei-Verlag. (Price 8.00 RM.).

DE SY, A. "*Siderurgie en Electrosiderurgie. Band 1.*" (Leerboek de Algemene Metallurgie, Siderurgie, Electrosiderurgie en Metallografie, Deel II). 8vo, pp. 454. Illustrated. Antwerpen, 1947 : Uitgeversmij N. V. Standaard-Boekhandel.

FEDERATION OF BRITISH INDUSTRIES. "*Scientific and Technical Research in British Industry.*" A Statistical Survey by the F.B.I. Industrial Research Secretariat. July, 1947. 8vo, pp. 26. London : The Federation. (Price 1*s.*).

GEOLOGICAL SURVEY OF GREAT BRITAIN. ENGLAND AND WALES. "*Geology of the Country around Whitney.*" (Explanation of Sheet 236). By L. Richardson, W. J. Arkell, and H. G. Dines. With contributions by C. G. T. Morison. 8vo, pp. 150. London, 1946 : H.M. Stationery Office. (Price 5*s.*).

GEOLOGICAL SURVEY OF GREAT BRITAIN. ENGLAND AND WALES. "*Geology of the Lizard and Meneage.*" (Explanation of Sheet 359.) By J. S. Flett and J. B. Hill. Second Edition by Sir J. S. Flett. 8vo, pp. xi + 208. London, 1946 : H.M. Stationery Office.

IRON AND STEEL INSTITUTE. Bibliographical Series No. 13. "*Bibliography on Wire (Manufacture, Treatment, and Properties) Including References to the Cold-Drawing of Bars.*" 8vo, pp. 146. London, 1947 : The Institute. (Price 2*s.* 6*d.* ; Members 1*s.*).

JIMENO, E. *El Problema de la Corrosion Metalica.* (Ministerio de Marina. Instituto Español de Oceanografia). 8vo, pp. 237. Madrid, 1947.

LUKENS STEEL CO. "*Steel Plates and Their Fabrication.*" Pp. 408. Coatesville, Pa. : The Company. (Price $5.00).